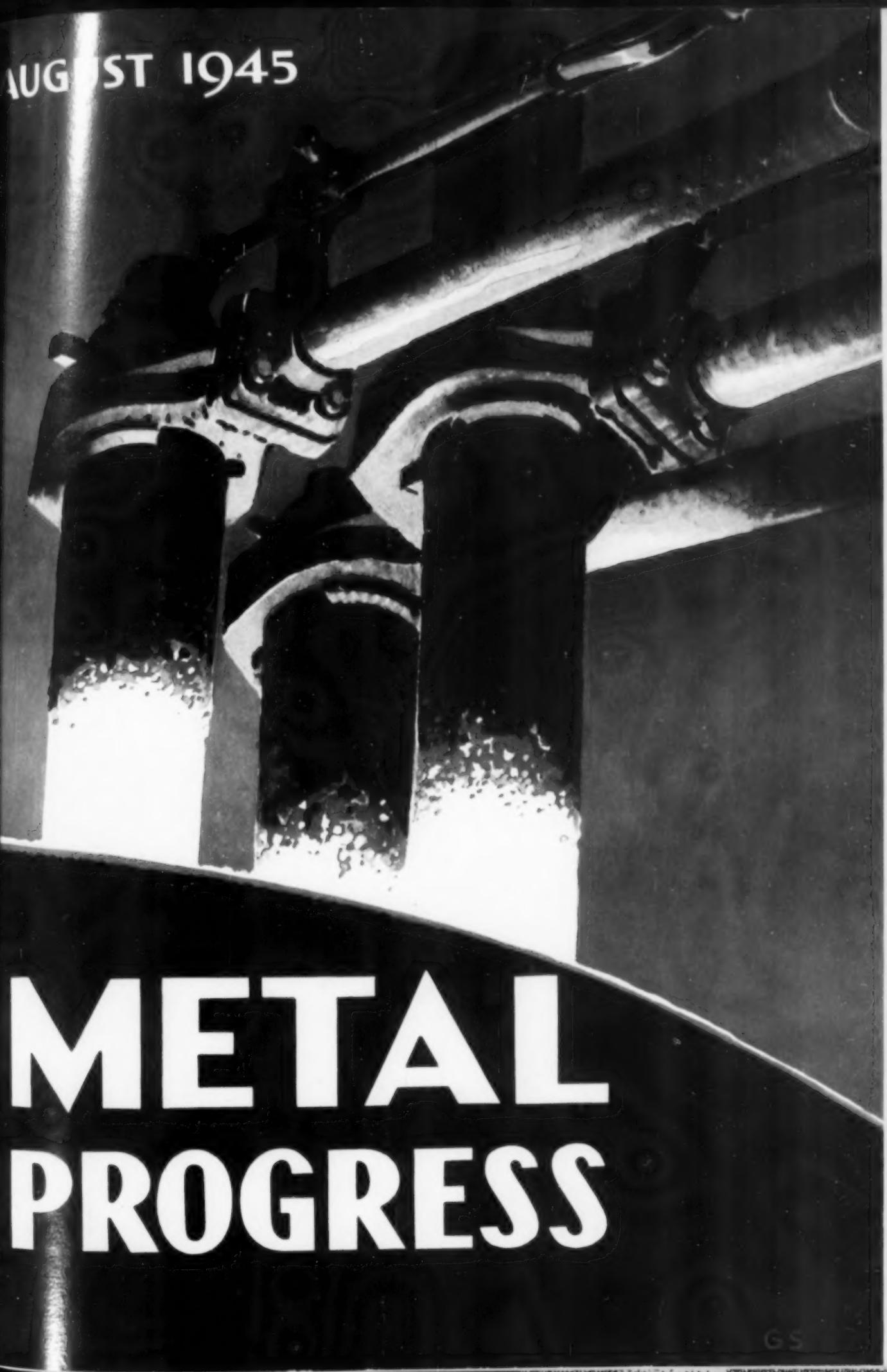
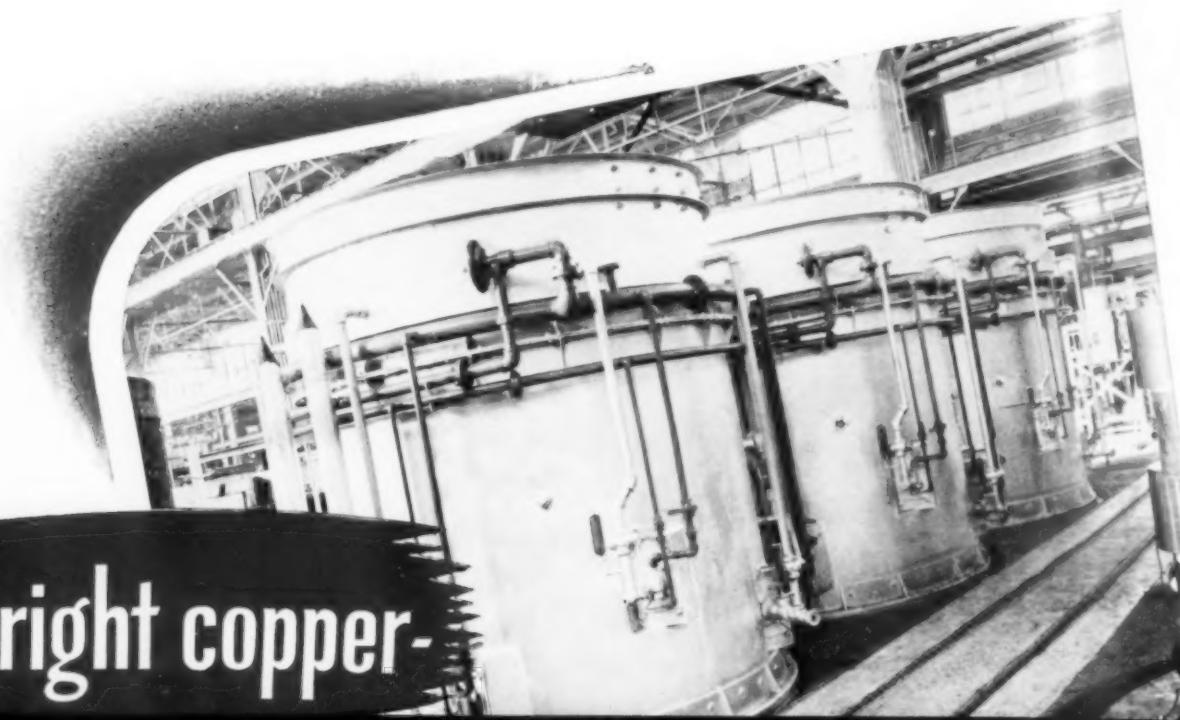


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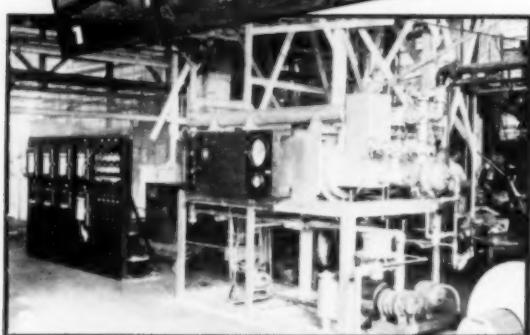
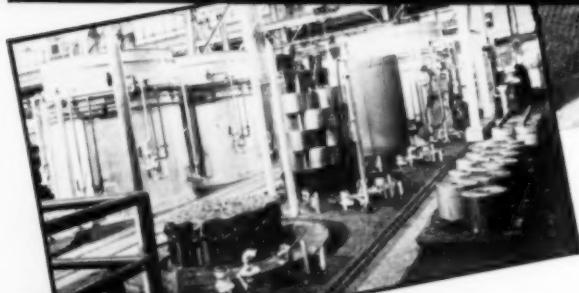


# METAL PROGRESS

GS



# bright copper- and clean brass heat processing



Above (1) View showing bases empty, loaded, and with inner cover. Each base is equipped with a recirculating fan for moving the hot gases. Each base has a capacity of 21 coils, or 12,600 pounds.

Below (2) DX Gas Preparation Unit in foreground. Two of these units supply the bases with the clean annealing atmosphere. Each has a capacity of 4,000 c.f.h.

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# 'Surface'

SURFACE COMBUSTION CORPORATION • TOLEDO 1, OHIO

August, 1945  
Volume 48, No. 2

# Metal Progress

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# Critical Points

By the Editor

TO DETROIT, to an exhibit of induction heating equipment sponsored by Detroit Edison Co., where nine manufacturers indulged in a friendly battle of frequencies, as it were—a little “heat treating show”, all of itself. The range ran from what originally was called “high frequencies” coming from motor-generator sets (960, 3000 and 9600 cycles per sec. are now more or less standardized), to frequencies up to the radio bands—say 450,000 cycles as a maximum—coming from oscillators of the spark-gap type and from electron tubes, and finally to ultra-high frequencies in the lower band of short wave senders, say 2 to 5 million cycles, generated in equipment akin to

radio broadcasting power

What frequency—  
three thousand  
or three million?

plants. (Above these frequencies is the field for heating dielectrics or non-conductors—such jobs as

the curing of plastics, the disinfection of packaged food, and similar non-metallurgical operations.) . . . . The best known applications of induction heating are, of course, for the surface hardening of steel parts, wherein heating currents of high frequency are induced in the surface layers. As a rough classification one can say that as the frequency goes up the penetration of the induced current (and the layer which can be directly and promptly heated) becomes thinner and thinner, the voltage (and required air gap between inductor coil and work) increases, and the cost of the necessary power goes up. An important practical consideration is that higher and higher frequencies are needed as the article becomes smaller and thinner—for example, 9600-cycle power cannot “flow” a tin deposit on tin plate (steel sheet about 0.010 in. thick) whereas the flux induced by sub-radio frequency of 200,000 cycles quickly heats the steel to the melting point of tin. Another practical matter

is that the electrical effect of hugging the surface means that irregular non-circular contours—especially small ones—are heated superficially by the higher ranges of frequencies; one of the exhibitors talked about thin “cases” on tiny gears that rival a dip in a cyanide bath, done with current oscillating at 2,000,000 cycles per sec. Surface hardening of cylinder bores with an internal inductor is an anomalous case; one might imagine that highest frequency would be necessary to keep the heating current on the *inside* of the cylinder rather than on the outside, but since the cylinders are usually fairly large a good deal of power is necessary to heat them rapidly else the hardened layer will deepen unduly by heat conduction. Hence the present desirability for

the powerful motor-generator sets of

*Tackling the hard jobs* 9600 cycles for heating inside surfaces. . . . The general rule that small items require higher frequency

does not rule out the possibility of deeper hardening of larger pieces if there is time enough and equipment available. The current efficiency will be somewhat less than maximum; nevertheless large pieces like 75-mm. armor piercing caps have been through hardened by very high frequency currents by keeping the current on for minutes and letting the generated heat soak inward and downward until all of the desired volume of steel is above the critical temperature. . . . Recent advances in the art have included the continuous heat treatment of long bars end-to-end with moderate frequencies, and thin walled tubing with moderate and higher frequencies. An extension of this idea might include the process annealing of sheet, strip and wire, but present impediments seem rather fundamental: Thin, fine metal requires very high frequencies in order to be heated at all, yet high frequencies mean intense surface effects and high heat gradients, surface to center, just opposite to the requirements for a uniform annealing effect. An interesting scheme for the finest material is to draw it through a close-fitting sleeve of graphite or heat resisting metal, which itself is heated by high frequency waves.

**WITNESSED** the casting of a 30-ton propeller of manganese bronze—something that Cramp Brass & Iron Foundries Division of Baldwin Locomotive Works can take in its stride, since production has been about four per day, of various sizes, ever since the war began. Molds are sectional—that is, there is a cope and drag for each of the four blades, and these form the hub when placed at accurate angles. Cement-sand mixtures are used; according to BONIFACE

**MILLER**, metallurgist, the principal advantage is that it is easier for poorly skilled workmen to make

**Propellers for Uncle Sam's ships** the molds (less danger of over-ramming); its principal disadvantage is that it takes three days to cure, or set, thus con-

gesting the floor. Casting troubles are prevented by very close moisture control in the sand-cement mix — modifying the percentage, in fact, with the season of the year. . . . Twenty-ton reverberatory furnaces melt the manganese bronze; it is transferred by preheated ladles to a large basin alongside the mold. After all dross has had a chance to rise, a stopper leading into the bottom sprue is opened. Four workmen keep watch through generous riser heads, one above each blade, and skim aside any dross with long-handled wooden rakes. Pouring is at 1820° F.; supplementary metal, sometimes as much as 3 tons, is added to risers at correct intervals. Blades are remarkably true to shape, and so the entire casting is balanced immediately after cleaning to determine the correct center for the shaft bore. After the hub is finish machined, the final balancing requires but a minimum of grinding on the blades.

**SEEMINGLY** the metallurgist must become an electrical as well as a chemical engineer if he is to understand the extraordinary machinery that is linked together to make an electrotinning line. Three of these lines are installed at the Fairfield works of Tennessee Coal, Iron and Railroad Co. (which, as all should know, is not in Tennessee but in Alabama — Birmingham, to be specific) and produce about two-thirds as much tin plate as the 41 pots in prior use for hot dipping, yet they occupy a new wing in the old building with hardly 10% its area. As a matter of fact, most of this space is occupied by the

**Electrotin plate for conservation** auxiliary devices; the actual tin-plating is done in one portion of each machine no bigger than an old fashioned tin pot, in which

the 30-in. wide strip is looped at high speed up and down a half-dozen times. (In fairness to the hot dip process it should be mentioned that the electrotin plate is coated with only  $\frac{1}{2}$  lb. of tin per base box — the industry's quaint way of saying 0.03 mil thick — whereas the common plating of the hot dip process is nearly three times as thick.) Ahead and beyond the central tinning unit are necessary auxiliaries such as reels for the cold rolled strip, a machine for welding the tail end of one coil to the leading edge of the next, magnetic hold-back to give necessary tension on strip in process, electrolytic

pickler and wash tanks. . . . After emerging from the tinning solution the strip is washed, then dried by steam, then the tin plating is "flowed" by electrical resistance, quenched in water, sprayed with chromic acid solution to give it a better tooth for lacquering, and oiled. It then goes through a power drive, a flying shear, an automatic kick-off for defectives, and the sheets finally land on a piler. Truly an amazing assembly, operating almost without human assistance, as indeed all such high speed mechanisms must. . . . Two devices are in fact superhuman.

**Superhuman devices for high speed processes** At the very start are a flying micrometer and an electric eye scanning for pin holes. When the gage goes off limits, or light leaks through, a peg is pushed

in on the rim of a "memory wheel". This is geared to the main drive, so when the portion of the strip that is defective reaches the end of the line, the memory wheel's pin has just reached a light beam that, through a photo cell and other mysterious robots, actuates the kick-out mechanism, rejecting the very sheet that has the defect plus the one ahead and the one behind, just to be sure. . . . Other non-understandable (to me) electrical shieldings are necessary to keep the currents put into the flying strip at three different places within the correct limits for their action. For example, the current used in the preliminary electrocleaning must not work forward to the tinning cell, neither must the heavy currents needed for flowing the tin deposit leak backward. (Melting the deposit, by the way, is a plain resistance job, the result of 5000 amperes of 90-volt current. To protect and flux the surface, the melting is done in a prepared atmosphere.)

**CANNOT REFRAIN** from praising the duplex process at Ensley plant (even though there is nothing new about it) for the unusual and truly spectacular operations are enough to catch the jaundiced eye. Here is a battery of three 25-ton converters, mid-length in a row of nine tilting openhearth furnaces; one or more of the vessels is always doing its pyrotechnic stuff. The humdrum facts are that the vessel oxidizes and slags off the silicon and manganese from the molten pig iron in about 5 min. and lowers the carbon to the desired point — either high for a "kicker", or dead soft in 7 to 9 min. more. The openhearts have deep baths that retain about half the steel at the end of each heat and all the finishing slag, highly basic and hot. Successive ladles of blown metal are then poured in and a violent boil mixes steel and slag and reduces the

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phosphorus almost instantly from 0.85 to around 0.04%. About 15 tons of the slag, carrying 10% or more  $P_2O_5$ , overflows the front doors and is replaced with fresh lime and mill scale. The heat is finished in the usual openhearth way, whereupon 140 tons of steel is drawn off from

*Tonnage from duplex steel-making*

below the slag line. All in a matter of 5 hr., tap to tap—35 heats a week are a commonplace, twice the number that can be produced in a stationary openhearth charged with hot iron, or three times the production from cold pig and scrap. The ground slag is an especially good fertilizer and conditioner for soils with acid reaction.... Aside from the specialized metallurgy, consider these favorable economics: Low openhearth plant overhead due to rapid production rates; low lining costs, due to silicon-free charge; no scrap required; credit from slag byproduct. To counterbalance these, in part at least, is the converting and transfer costs, both operating and overhead, and the losses in iron oxide. The salable slag carries the equivalent of 12% metallic iron, and therefore about 1 ton of iron goes out in slag for every 100 tons of ingot cast. However, that is not much more than an ordinary openhearth operation would lose. The larger loss, about 5 tons of iron per 100 tons of ingot, is in converter spillage and slag; the latter is an iron silicate of low value in this district where the ores (even after most canny mixing) are on the siliceous side of neutral.

*Centrifugal casting of magnesium retorts and ship shafts*

FASCINATED with operations at the American Cast Iron Pipe Co. in Birmingham, where many years of experience with centrifugal casting of iron pipe in 16-ft. lengths up to 48 in. diameter have been turned to the manufacture of similar tubes of alloy iron, carbon and alloy steel, and tubes of chromium-nickel-iron alloy (retorts for magnesium distillation). Short items like bogie wheel hubs for tanks are oxy-acetylene cut from the 16-ft. lengths; long items like ship shafting are often double-lengthed and flanges attached by welding. An unusual example of alloy con-

servations was also noted: Die pot liners for shell forging are made of Ni-Mo-Cr cast iron; rough castings are sent to the shell forging plant in return for the turnings and the worn liners; the actual metal has

already made many round trips through the cupola, and it only requires sweetening with a little chromium each time. In the shell forging shop the average production has crept up to 1200

shells per pot.... In company with CHARLES DONOHO, metallurgist, spent much time in the department making gas engine cylinder liners. Since these have a stubby flange on one end they are cast individually with axis horizontal, spinning at such speed that the centrifugal force is 60 times gravity. The cast iron mold is about five times as thick as the casting's wall in order to chill it at the proper rate, and is bare except for a spray of refractory slurry. The flanged end is closed by a stopper plate, keyed in. Steel is 4137 (1.00% chromium, 0.20% molybdenum), melted in one-ton acid electric furnaces, killed with silicon and manganese—no aluminum. A correct amount is weighed into a hand ladle, and then poured rapidly into a small pouring basin; a submerged spout leading into the spinning mold holds back any floating slag, and its diameter governs the rate at which the steel is cast. Within a minute the metal, originally at 2850° F., has cooled below 2600° and frozen; the mold and contents are lifted out of the cradle and replaced with an empty one. Six machines, each spinning four molds simultaneously, produce 800 of these 35-lb. castings in 8 hr.... JAMES MACKENZIE, chief metallurgist, points out that the ancient

#### *Cast vs. forged cylinder liners*

argument about whether a test bar represents the properties of a casting is resolved by cutting a block right out of the cylinder wall. These liners, as cast, have a uniformly textured Widmanstätten microstructure which is readily recrystallized into an equiaxed structure, about grain size 6, by heating to 1600° F. for 1½ hr., air-cooling, and drawing at 1225°. Hardness is then Brinell 200 or less, suitable for machining; about  $\frac{1}{4}$  in. is taken from the inside and  $\frac{1}{8}$  from the outside, thus removing almost half of the original wall. Every cylinder is magnafluxed before shipment; internal surfaces must be absolutely sound and perfect. Not only does it save freight to keep defectives at home, but the foundry can immediately correct conditions that are making trouble. Final heat treatment in the engine plant consists of an oil quench from 1550°, and a draw at 1050° F.; structure is then a uniform fine sorbite with mere traces of excess ferrite; hardness is C-32 to 34. Forged barrels of the same analysis have approximately the same physical properties except that the average elongation is 15% when tested longitudinally, but only half as much transversely. MACKENZIE says also that the real test of these cylinders is the 150-hr. proof run of completed engine without sign of scuffing or wear. (2)

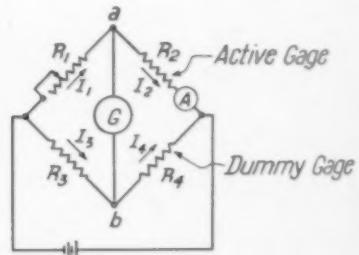
By Given Brewer  
Structures Engineer  
Consolidated Vultee Aircraft Corp.  
Vultee Field, Cal.

# Interpreting and Recording Data From SR-4 Gages

A BRIEF ACCOUNT of electric strain gages, that have proven so useful in the recent studies of complex structures such as airframes, ships, engines and heavy ordnance, was printed in the last issue of *Metal Progress*. Having given some information on their construction and method of attachment, on the theory of their action and response to temperature and strain, the present article will describe some simple assemblages of electrical equipment necessary to measure and record their indications.

Since movement (strain) in the metal at the spot on which the gage is cemented is reflected by a change in electrical resistivity of the wire of which the gage is made, adequate electrical equipment is required to measure changes in the flow of an imposed current as the resistance of the circuit changes. The type of such equipment varies within wide limits depending on the speed at which the measurements are to be made, on the distance between the object being tested and the measuring instruments, on whether the test is being made on a stationary or a moving object, and on whether individual measurements or autographic records are to be secured. A full discussion of the various successful wiring and instrumental diagrams would be almost endless; what will be attempted here will be merely a description of a simple circuit for a structure which can be loaded by steps and the loads held constant for any time required to make the individual readings.

Such simple equipment would usually comprise a resistance measuring device based on the Wheatstone bridge principle. The circuit for the fundamental type is shown in Fig. 1. The resistances shown at  $R_2$  and  $R_4$  are strain gages and are cemented to the active structure and the dummy structure respectively (the dummy structure being used for temperature compensation, as outlined in the preceding article). The investigator usually utilizes the null method, in which the bridge is brought into balance by varying  $R_1$  or  $R_3$ , balance being obtained when the galvanometer  $G$  reads zero voltage — that is, no current is flowing between the bridge-ends  $a$  and  $b$ .



Let  $I$  = Current; Amperes

$V$  = Voltage; Between  $a$  and  $b$

$\beta$  = Temp. Coef.; Ohms/Ohm/°C.

$\Delta t$  = Temp. Change; °C.

$R$  = Resistance; Ohms

If  $V=0$ , the Bridge is  
Balanced and:

$$I_1 = I_2; I_3 = I_4$$

$$R_1 I_1 = R_3 I_3; R_2 I_2 = R_4 I_4$$

$$\therefore \frac{R_1}{R_2} = \frac{R_3}{R_4} \quad (1)$$

Fig. 1—Diagram of Simplest Circuit Utilizing the Wheatstone Bridge

From equation (1) in the diagram it can be seen that if the active and dummy structures undergo an equal change of temperature equal to  $\Delta t$ , then, due to temperature alone:

$$R_{2(t)} = R_2 + \Delta t \beta R_2$$
$$R_{4(t)} = R_4 + \Delta t \beta R_4$$

(Notation is as in Fig. 1. For a definition of the temperature coefficient of the strain gage  $\beta$  and its numerical values, see the text and Table I in the article last month.)

Therefore:  $\frac{R_1}{R_2 + \Delta t \beta R_2} = \frac{R_3}{R_4 + \Delta t \beta R_4}$   
 or  $\frac{R_1}{R_2(1 + \Delta t \beta)} = \frac{R_3}{R_4(1 + \Delta t \beta)}$  (2)

and as  $\Delta t \beta$  is the same for the active and the dummy gage the bridge remains balanced when

$$\frac{R_1}{R_2} = \frac{R_3}{R_4}$$

which is the same as equation (1) in Fig. 1.

Presuming careful technique and placement of the dummy gage it can be seen from equation (2) that nothing but changes in resistance of the active gage  $R_2$  can throw the bridge out of balance. As the structure is loaded the resistance of the active gage changes in proportion to the surface strains of the structure along the axis of the gage. The voltmeter (galvanometer  $G$ ) will then indicate a difference of potential between bridge-ends  $a$  and  $b$ , and it must be brought back into balance by changing the resistance of  $R_1$  in accordance with the equation (1) until the voltmeter again reads zero.

If a temperature change  $\Delta t$  has taken place during the time of loading the structure, the equation for balance, including resistance changes due to loads and resistance changes due to

thermal effects, is as shown in equation (3):

$$\frac{R_1 + \Delta R_1}{R_2 + \Delta R_2 + (K\Delta t + S_a \Delta t)R_2} = \frac{R_3}{R_4 (K\Delta t + S_a \Delta t)} \quad (3)$$

Since  $(K\Delta t + S_a \Delta t) = \beta \Delta t$ , as shown in the first article, it also can be shown that

$(R_2 + \Delta R_2) (1 + \Delta t \beta) = R_2 + \Delta R_2 + \beta \Delta t R_2 + \Delta t \beta \Delta R_2$   
 where  $\Delta R_1$  is the change in resistance of  $R_1$  necessary to balance the bridge after it has been unbalanced by a change  $\Delta R_2$  in the active gage due to strain in the structure.

Equation (3) then becomes (neglecting the second order differential term  $\Delta t \beta \Delta R_2$ ):

$$\frac{R_1 + \Delta R_1}{(R_2 + \Delta R_2) (1 + \Delta t \beta)} = \frac{R_3}{R_4 (1 + \Delta t \beta)}$$

and  $\frac{(R_1 + \Delta R_1) R_4}{R_2 \times R_3} = \frac{R_2 + \Delta R_2}{R_2} = 1 + \frac{\Delta R_2}{R_2}$

or  $\frac{(R_1 + \Delta R_1) R_2}{R_2 \times R_1} = 1 + \frac{\Delta R_2}{R_2}$

$$1 + \frac{\Delta R_1}{R_1} = 1 + \frac{\Delta R_2}{R_2}$$

$$\therefore \frac{\Delta R_1}{R_1} = \frac{\Delta R_2}{R_2} \quad (4)$$

Since  $\Delta R_1$  and  $R_1$  are known,  $\frac{\Delta R_2}{R_2}$  may be calculated from (4).

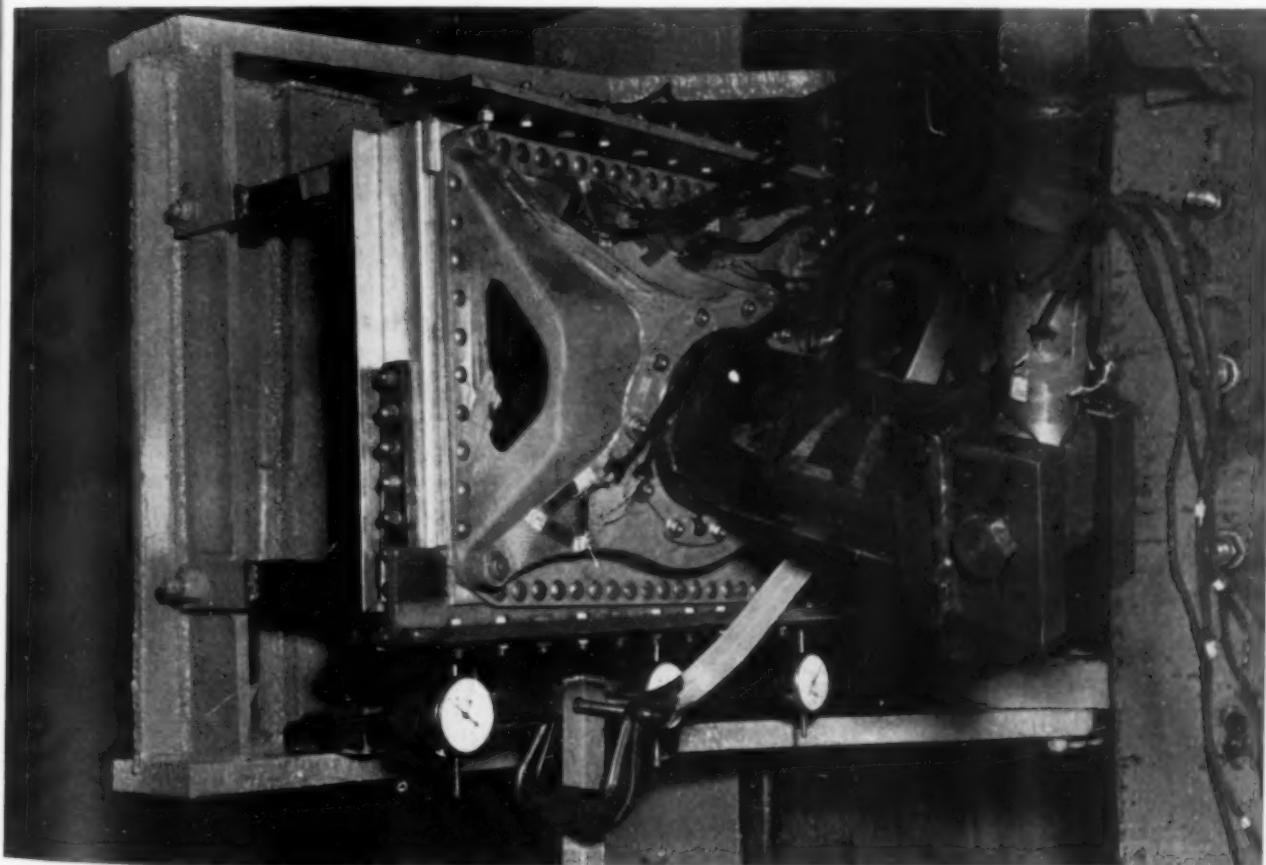


Fig. 2—Strain Gage Rosettes (Triangles) Placed on Webs and Body of Heavy Fitting to Determine the

Effect of Refining the Design. Note the tool marks alongside upper right web where structure has been lightened

In the former article it was shown that the strain sensitivity factor  $S$  of the gage used (i.e. +2.38 for Copel wire or +2.82 for iso-elastic wire) is equal to

$$S = \frac{\Delta R}{R} - \frac{\Delta L}{L}$$

But since the axial strain  $\epsilon_a$  in the structure along the gage axis equals  $\frac{\Delta L}{L}$ , then

$$\epsilon_a = \frac{\Delta R_2}{R_2 \cdot S} = \frac{\Delta R_1}{R_1 \cdot S} \quad (5)$$

Using the hook-up of Fig. 1 it is very easy to measure the axial strain in a tension member, for example. After the gage has been cemented in place, the connections made, the circuit balanced, and the resistance  $R_1$  noted, a load is then applied, the electrical circuit again balanced and the change in resistance  $\Delta R_1$  noted. The axial strain is then figured from equation (5) and the stress by multiplying the strain by the modulus of elasticity of the metal being tested — that is,  $f_a = E \cdot \epsilon_a$ .

Using the null balance method and reversing the battery in Fig. 1 to isolate thermo-electric currents arising from dissimilar metal contacts in the leads, it is possible to measure strains of 0.000,001 in./in., corresponding to a load of 30 psi. in steel or 10 psi. in aluminum alloys. An additional advantage of the null method is that it is independent of fluctuations in the battery voltage.

If a voltmeter is used in the bridge and it does not draw appreciable current (such as an electrostatic voltmeter or a galvanometer with very high internal resistance) the strain may be found by the relationship:

$$\epsilon_a = \frac{2V}{I_2 \times R_2 \times S} \quad (6)$$

where:  $V$  = voltage change due to unbalance in circuit

$S$  = gage strain sensitivity factor  
= 2.04 for A-1 gage

$I_2$  = ammeter reading, amperes  
 $R_2$  = resistance of gage  
(120 ohms for A-1 gage)

$\epsilon_a$  = axial strain along axis of gage in./in.

Utilizing this alternative method and equation (6) strains may be determined more quickly than by the null method, but with less accuracy. Both of the methods described in the foregoing paragraphs are of course suitable only for measuring static loads that can be held constant long enough to achieve balance and make readings.

Equipment available commercially for the

measurement of static strains\* includes a 48-gage recorder that automatically records static strains from 48 gages in sequence on a polar coordinate chart. This machine operates by balancing a Wheatstone bridge, but employs a 1000-cycle carrier wave in place of the direct current (from battery or other source) of the bridge illustrated in Fig. 1. As the active gages are switched in to the recorder its strain is determined automatically by the machine and a pen records this value on the chart, whereupon the chart moves 1/48 of a revolution and the next gage is switched into the circuit. A single dummy gage is usually employed

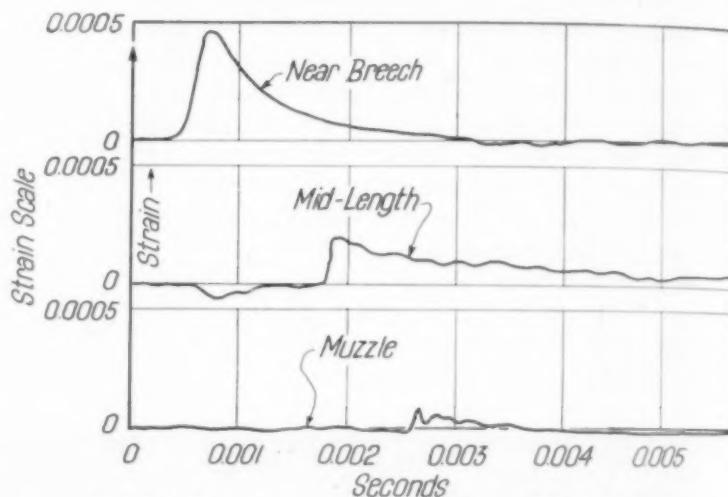


Fig. 3—Strains at Breech, Muzzle and Mid-Length of Shotgun Barrel When Gun Is Fired (de Forest)

and this is automatically switched into the 48-point recorder from time to time, thereby compensating for temperature changes.

#### Measurement of Dynamic Strains

As the inertia of the electric strain gage is negligible its resistance change follows the rate of change of axial strain of the stressed structure without measurable lag. In fact the strains along a shotgun barrel have been recorded with the passage of the charge (Fig. 3). Similar studies on machine guns and cannon, as well as armor plate, have uncovered important data for designers of ordnance.

\*Manufacturers of strain gage equipment include the following:

Baldwin Locomotive Works, Baldwin Southwark Division, Philadelphia, Pa.

William Miller Co., Pasadena, Cal.

Consolidated Engineering Co., Pasadena, Cal.

Hathaway Instrument Co., Denver, Colo.

Heiland Instrument Co., Denver, Colo.

General Electric Co., Schenectady, N. Y.

Westinghouse Electric Corp., Pittsburgh, Pa.

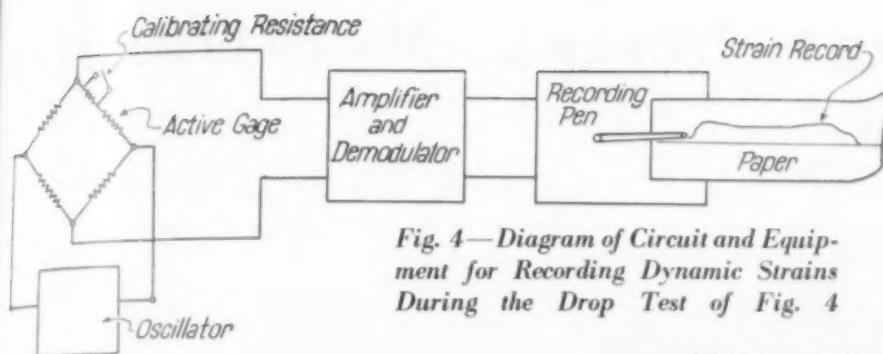


Fig. 4—Diagram of Circuit and Equipment for Recording Dynamic Strains During the Drop Test of Fig. 4

To record dynamic strains it is necessary to employ electronic rather than mechanical devices capable of following the high strain rates associated with dynamic loadings. The dynamic equipment is similar to the static equipment in that both are based on the measurement of voltage unbalance in a Wheatstone bridge. The change in voltage potential between points *a* and *b* (Fig. 1) is amplified by electron tubes or other electronic devices.

For moderate frequencies, up to 120 cycles per sec., pens actuated by piezo-electric crystals are able to record the dynamic strains on moving paper. An example is shown in Fig. 4. Mirror type multi-channel recording oscilloscopes will function satisfactorily at frequencies up to 2000

weight and the weight of the drop test jig. These stresses corresponded to one load factor (or  $1g$  for gravity) on the gear. The gear was then raised until the tire was 15 in. above the floor, whereupon the trigger was pulled and the strut and gear dropped. Stresses on the strut

resulting from the deceleration of the jig weight by tire compression, plus the reaction of the hydraulic fluid passing through the orifice in the landing gear strut when the entire combination hit the floor, were measured and divided by the initial stress value found for one load factor, the quotient yielding the load factor after several durations of time interval from the moment of contact with the ground (about 0.007 sec. after the time base started).

The resultant curve is Fig. 5. During the test just described a space-time record of the mass travel of the strut was also taken by cable-driven pencils which traced a line on a drum driven by synchronous motor. Load factor vs. time was made by graphically determining the second derivative of the space-time curve and adding 1 load factor to this deceleration curve.

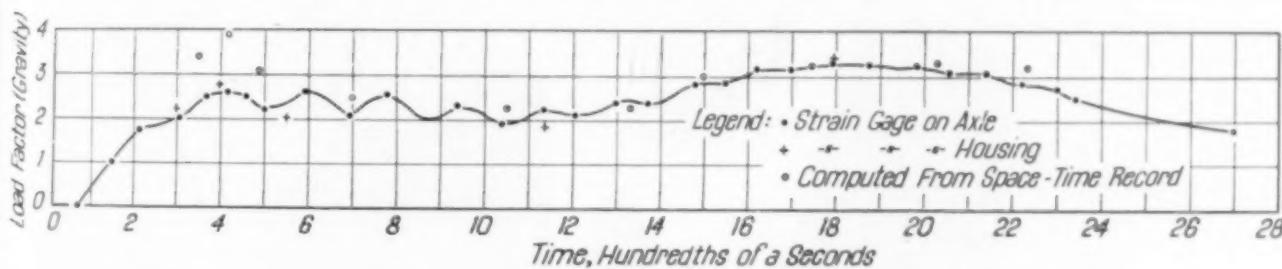


Fig. 5—Loads Endured by a Certain Landing Gear for 0.27 Sec. After Enduring a Drop Test

cycles per sec., the record being made on photographic paper or film. Above these frequencies cathode ray oscilloscopic devices must be employed, the record also being made photographically. Some success has also been attained in the recording of aircraft flight strains on magnetic steel wire, impressing thereon a record which may be played back and analyzed later on the ground. Strains encountered in flight have also been impressed on radio carrier waves which are received at a ground station and analyzed at once.

The use of the piezo-electric actuated pen in recording landing gear stresses during drop test is illustrated in Fig. 5. The static stress at two points on the strut was recorded with the landing gear on the ground supporting its own

Several such determinations are plotted on Fig. 5 by small circles. Comparing the results of the two methods for determining load factor it is apparent that both the strain gage and the space-time derivative method yield valid results; however, the strain gage record is obtained in a fraction of the time required by the graphical analysis of the space-time record. The discrepancy between the two methods 0.04 sec. after the beginning of the stroke may be attributed to elastic over-travel in the cable-pen recording system.

Problems associated with the recording and computation of strains in two-dimensional stress fields and in buckled panels, the use of strain rosettes and the correction of results necessitated by the "cross sensitivity" of the gages will be discussed in later issues of *Metal Progress*.

# An Eminent Living Metallurgist



*Joseph Donald Hanawalt*

**Expert in Magnesium and Its Alloys**

OF THE MANY INDUSTRIES which have enjoyed unusual growth or development during the course of World War II, none has captured the interest of the American people more than the magnesium industry in its rise from obscurity to prominence, multiplying production capacity by almost one hundred times. This rise from a little more than 6,000,000 lb. in 1939 to nearly 600,000,000 lb. in 1944 attended the first real awakening of the United States government and American industry to the potentialities of this metal.

Cradled in World War I when German supplies were cut off, the American magnesium industry barely survived the interim of peace. After the military demand for magnesium in bombs, flares, and other pyrotechnics had subsided in 1918 only the faith of two American concerns, the Dow Chemical Co. and the American Magnesium Corp., in the worth of the metal as a structural material kept production alive. Shortly after the close of the first war all but these two producers dropped out, and in 1927 American Magnesium Corp. ceased production of the metal and concentrated on its fabrication, leaving Dow as the sole American producer of magnesium for the 13 years prior to 1941.

There is little doubt that this country's ability to out-produce the world in magnesium today is to be credited to the knowledge acquired between wars by these two independent firms. Recent production figures provide ready evidence of the ability in America to make the *metal*. While the ability in America to produce the highest quality magnesium alloy *structures* is not as obvious, it is nevertheless equally existent. If a comparison would be made of the American metallurgical development of alloys and of the technical processing of magnesium structural parts with practices of foreign countries during the current war, it would show that in these fields America is well in advance.

We are sometimes prone to speak of "companies" being responsible for certain achievements, forgetful of the fact that in actuality a company is but a legal entity and that it is to some of the human personalities involved therein that the real credit is due. In the case of magnesium at Dow Chemical Co., many men were engaged with continually increasing intensity in the decade preceding World War II. Their developments of production and fabrication tech-

niques enabled magnesium in all its aspects to gear itself to war tempo in a few short months. Principal among these men is JOE HANAWALT, associated with Dow since 1931 and today director of the company's extensive metallurgical research.

JOSEPH DONALD HANAWALT was born in Royersford, Pa., near historic Valley Forge, in 1902. His family moved to Akron, Ohio, where he received his early education. He first became interested in physics while attending Oberlin College, where he took his Bachelor's degree in 1924. After graduation, he went to the University of Wisconsin, where he served as assistant and instructor in physics for five years, receiving his Doctor's degree in 1929. He then spent two years as National Research Fellow of the Rockefeller Foundation, studying X-ray absorption spectra at the University of Michigan and the University of Groningen.

At this time the Dow Chemical Co. was in the process of expanding its work in the spectroscopy and X-ray departments, and it was thus that Dr. HANAWALT became associated with the firm. Under his direction the work in these fields was expanded, and Dow's leadership and pioneering developments in the practical applications of physical analytical methods in the chemical industry are well known to chemists and physicists. Similar rapid methods of accurate analysis are today more common in metallurgical industry, even getting into the furnace rooms and melting shops so the melters may have continuous knowledge of the refining process. A particularly successful development of these first years, now in wide use, was a method of indexing X-ray diffraction patterns so they may be used for identifying unknown substances or mixtures of unknowns. At the present time, one of the unique features of the Dow Metallurgical Dept. is this highly developed division of physical techniques — X-ray, spectroscopy, electron diffraction, mass spectroscopy — and their application to metallurgical problems.

These things have contributed to the transformation of many phases of the magnesium industry. Ten years ago the pure metal was handled on small scale with expensive techniques, and its alloys were primarily characterized by lightness alone. Now there are magnesium alloys having excellent weldability, formability, strength, and corrosion resistance. Under HANAWALT's leadership Dow research has changed the processing steps so fundamentally that magnesium has left the rule-of-thumb stage and entered a new era. An achievement considered unattainable has been the development of a reverberatory fur-

nace for large scale, safe, and economical melting of the metal. Continuous casting has been introduced into mass production; pumping and piping of molten magnesium have been used to advantage in various processes. HANAWALT has a vision of plants for alloying, refining, and casting of magnesium which are as clean and as pleasant to work in as a bakery. He sees cost advantages for magnesium in its light weight, low melting point and adaptability for processing in iron equipment, which no other metal possesses.

In his approach to industrial research and in his conduct of the metallurgical department, he has patterned his philosophy closely after that of his chief, WILLARD H. DOW. JOE HANAWALT does not sit as an arbitrary head of his department. He believes in giving responsibility and in developing the individual through this method. He gives a man adequate facilities so he can explore lots of territory in the pursuit of his project. He encourages his men to utilize to the fullest degree modern theory and techniques as applied to their work, but he is wary of decisions based on theory which are unsupported by experiment. He encourages the Edisonian approach and suggests that one should make use of a theory to suggest new experiments, but should not let a theory go very far in ruling out an experiment. He believes in the value of the competitive system to invigorate an organization and in the importance of each man receiving the rewards of his efforts. He is convinced that a unified group will accomplish many things which no man alone can attain.

Dr. HANAWALT is a member of the American Society for Metals, American Institute of Mining & Metallurgical Engineers, American Society for Testing Materials, American Foundrymen's Association, American Physical Society, American Chemical Society, The Electrochemical Society and American Society for X-ray and Electron Diffraction. He serves on committees of several of these organizations, as well as on the National Defense Research Council's committees on magnesium projects. He is a member of the Board of Directors of the Detroit Section of A.I.M.E. and is a past-chairman of the local section of the American Chemical Society. In 1943 he received the Institute of Metals' Annual Award for his outstanding studies of the corrosion resistance of magnesium and its alloys.

Outside of his work, Dr. HANAWALT is a ski enthusiast. He joins in the activities of the Rotary and Torch Clubs and serves as a trustee of his church. He likes music and is a member of the Dow chorus.

SAMUEL CROWTHER, 3rd.

## Non-Destructive Spectrographic Sampling

WITH the recent pronounced trend toward spectrographic analysis for metals used in industry, it has developed that one of the major problems is to obtain satisfactory samples. The desirable qualities of a sample are: (a) That it be in a form that provides satisfactory analytical accuracy, (b) that excessive time is not consumed in its preparation, (c) that the number of parts scrapped in obtaining the samples is held to a minimum, and (d) that they may be handled conveniently in the laboratory.

As far as the actual analytical work is concerned, samples in the form of pins are more conducive to accuracy and are most easily handled in the sparking procedure. Samples of this type are usually  $\frac{7}{32}$  or  $\frac{1}{4}$  in. round, with ends ground flat, round, or with a rather flat cone. If one is controlling a melting process, the preparation of these pins is rather simple since they may be cast in a mold or merely sucked up into a glass tube. However, if a forging, casting or bar stock has to be checked, the work of getting a sample may become rather tedious. For this reason, the technique of sparking a flat surface was developed and is now being used in some laboratories, and a satisfactory degree of accuracy is reported.

We used this method for some time, but there were always certain ranges of some of the elements that did not lie within the region of good reproducibility. In view of the damage usually done to the original part in the

By S. L. Widrig  
Chief Metallurgist  
and F. W. Lutz  
Research Engineer  
Toledo, Ohio

preparation of a rod-shaped sample, some method of sparking a flat is most desirable; however, in addition to the limitation of accuracy, large parts are cumbersome to arrange on the spark stand.

To overcome these difficulties, samples are now being prepared in this laboratory by removing a chip or fragment from the part to be analyzed, and welding it to the end of a  $\frac{1}{4}$ -in. round pin of mild steel. The chip is then ground to the circular dimension of the pin and a cone is ground on the end of it, all as shown in Fig. 1 and 2, pages 278 and 279. The chip is small enough that it does not involve scrapping the piece. The time required is only slightly more than that for preparing a flat surface, but is not excessive in view of the other advantages.

At the present time, when the labor supply is limited, it is a distinct asset to have small samples of this kind, rather than heavy pieces that cannot ordinarily be handled by a girl spectrophotographer.

Tests have shown that any uniform welding technique that produces a reasonably good bond will have no serious effect upon the analysis. The important functions of the weld are to provide sufficient electrical conductivity to the end of the pin, and sufficient heat conductivity away from the end. A bond of any kind will provide the

Table I lists analyses on identical pieces of steel under three different conditions of sample preparation. In the first, a solid pin of the steel was used and no welding was involved. In the second, a small chip was welded according to the usual technique. In the third, the chip was purposely very severely overheated white-hot in welding, a condition far more severe than ever encountered.

These tests show that the degree of heat during welding does influence the accuracy somewhat; however, these results

Table I—Results on Five Samples Made in Each of Three Ways

	SOLID PIN	WELDED PIN	OVERHEATED WELD
4620 Steel			
Mn	0.55 to 0.58; Av. 0.560	0.56 to 0.61; Av. 0.588	0.59 to 0.60; Av. 0.598
Ni	1.72 to 1.77; Av. 1.748	1.69 to 1.74; Av. 1.710	1.65 to 1.70; Av. 1.678
Cr	0.54 to 0.55; Av. 0.548	0.58 to 0.62; Av. 0.602	0.54 to 0.56; Av. 0.550
Mo	0.23 to 0.24; Av. 0.232	0.24 to 0.25; Av. 0.242	0.23 to 0.24; Av. 0.232
Si	0.28 to 0.29; Av. 0.278	0.30 to 0.32; Av. 0.312	0.29 to 0.32; Av. 0.310
8620 Steel			
Mn	0.89 to 0.92; Av. 0.900	0.91 to 0.94; Av. 0.927	0.94 to 0.95; Av. 0.948
Ni	0.48 to 0.51; Av. 0.490	0.46 to 0.47; Av. 0.464	0.48 to 0.50; Av. 0.488
Cr	0.49 to 0.54; Av. 0.522	0.54 to 0.58; Av. 0.564	0.51 to 0.53; Av. 0.516
Mo	0.21 to 0.22; Av. 0.214	0.22	0.21 to 0.23; Av. 0.218
Si	0.22 to 0.27; Av. 0.250	0.26 to 0.27; Av. 0.268	0.27 to 0.29; Av. 0.280

represent extreme conditions. If a reasonable amount of care is used in welding, so the maximum temperature does not vary over too wide a range, the analytical results stay within acceptable limits of variation. If only the grade of steel is to be identified, any type of weld would give suitable accuracy.

One explanation for this variation is that the elements become oxidized. Oxides of some elements reduce the line intensities while the oxides of others increase them. This difference in effect can be seen by comparing the results for manganese and chromium.

Table II illustrates that these variations are not excessive if a reasonable amount of care is used in the welding operation. In this run, eight chips were cut from one piece of 8620 steel and welded separately to eight pins. These results are representative of the reproducibility found in the routine work.

Chips that may be used vary in thickness from  $\frac{1}{8}$  to  $\frac{3}{16}$  in. and need be only slightly larger in area than the  $\frac{1}{4}$ -in. diameter pin to which they are welded. After trying a number of methods, it was found that the chip could be easily pushed off a corner of the piece to be analyzed by a chisel mounted in a small hydraulic press (10 tons capacity). Since most of the

Table II—Variations in Eight Samples Made From a Single 8620 Forging

ELEMENT	MINIMUM	MAXIMUM	AVERAGE
Manganese	0.80	0.84	0.820
Nickel	0.49	0.52	0.504
Chromium	0.50	0.52	0.512
Molybdenum	0.15	0.16	0.154
Silicon	0.21	0.23	0.220

necessary electrical conductivity, and it was found that an air space extending approximately one-third through the weld was necessary before the heat conductivity was limited to where it influenced the analytical results.

The welding temperature was found to be the most important variable. To illustrate this,

materials handled in this laboratory are not heat treated, this scheme works quite well and requires only a few seconds. The sample may also be removed with a high speed cut-off wheel if the part is very small or has been hardened. The chip is then welded to a pin of 1020 steel in a resistance welder (Fig. 1), and ground (Fig. 2).

It should be pointed out that only one chip is prepared in this way from each sample and an opposing electrode of relatively pure iron is used with it in the spectrograph. Such a practice does not seem to influence appreciably the degree of reproducibility. It might be expected that a higher degree of accuracy could be obtained if both pins were made up of the sample, since twice as much of the elements being measured would be vaporized. Any such benefit seems to be rather slight and does not warrant the additional work of preparing two samples.

### Useful for Thin Sheets

It is interesting to note that such a sampling method also provides a suitable means of analysis for thin sheets of metal. These have always presented a considerable problem in that the heat dissipation is inadequate and the temperature of the surface being excited becomes rather unpredictable. By preparing the specimen as indicated in this article, spectrographic analyses check chemical analyses as closely as they do for any other size or shape. The only limitation to the thinness of the sheets that may be handled in this way seems to be that they must have sufficient mass to be welded and ground to the proper shape.

(Some good work has been done with pins having flat ends. This suggests the possibility

of analyzing extremely thin sheets by the above method, since only enough grinding would be necessary to present a clean surface for excitation.)

The spectrograph used in this work is a "Bausch & Lomb Large-Littrow" type with quartz optics. Since it is used at various times for both flat surface specimens and pins, it is mounted on three rollers so that it can be moved longitudinally by a hand wheel to line up the slit with the two different sources. In addition, four lenses with different focal lengths are mounted in a turret before the slit so that the source-to-slit separation can be varied over a wide range.

The microphotometer and the spark source were both built by the Engineering Research Department, University of Michigan. All of the exposure sequences are controlled automatically, including racking the plate down to the next position after one exposure has been finished. All points of access to the high voltage circuit are protected with safety switches and solenoid locks, and these, in turn, are connected with a switch that short circuits the main busses when the source is not operating. This arrangement makes it impossible

for the operator to come in contact with any part carrying high voltage during its operation, or to start the discharge if all of the doors have not been properly closed and locked.

The synchronous interrupter at the source, which is the device usually furnished with this spectrograph, has been replaced by an air gap interrupter similar to the one described by R. A. Wolfe and E. J. Jemal in *A.S.T.M. Bulletin No. 129*, August 1944, and later by R. G. Fowler and R. A. Wolfe in *Journal of the Optical Society of America*, Vol. 35, February 1945.

Since this is one of the first industrial appli-

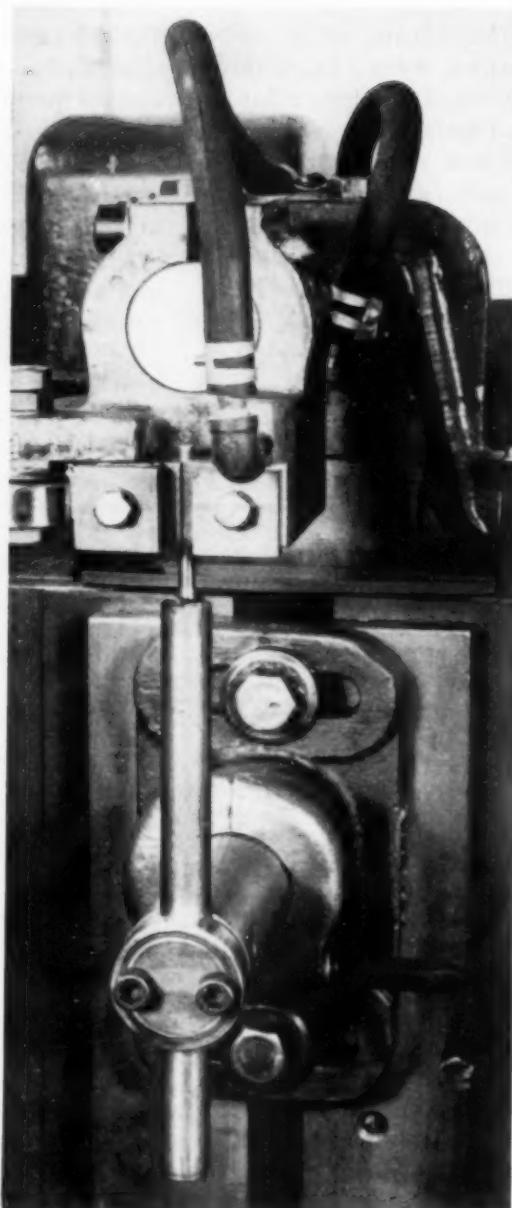


Fig. 1—Spot Welder for Preparing Spectrographic Pencils. The square fragment cut from the part to be analyzed is placed on top the lower electrode and a  $\frac{1}{4}$ -in. round of 1020 steel is gripped between the upper electrode jaws

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cations of this type of interrupter, it seems worthwhile to mention that it has been found a distinct improvement over the mechanical one previously used. The principle of its operation is obvious from the photograph reproduced in Fig. 3. It is placed in the electrical circuit in the same position as a synchronous interrupter and the air jet blowing between the two tungsten electrodes sweeps away the ionized air and a new discharge occurs as soon as the condenser has acquired sufficient charge to jump the gap again. Therefore, the number of discharges per cycle is a function of the rate at which power is supplied to the condenser and the peak voltage is governed largely by the separation between the tungsten electrodes.

By this device it is possible to have 5 or 10 (or even as high as 30) discharges of equal magnitude within a half cycle of voltage from whatever power supply is being used. This increase in number appears to excite a prepared surface more thoroughly.

Preliminary tests indicate that such an increased number of discharges might also improve the analytical accuracy obtainable with flat surface specimens. It is generally believed that the principal fault in flat surface sparking technique is the inability to control the discharge over a given area of sample. For this reason,

one microscopically small part of the area might be considered to receive the spark only once in a number of discharges and the frequency with which it does receive the charge may be irregular, so that vaporization from that particular area is not uniform. If all other factors remain constant and the number of discharges within a given

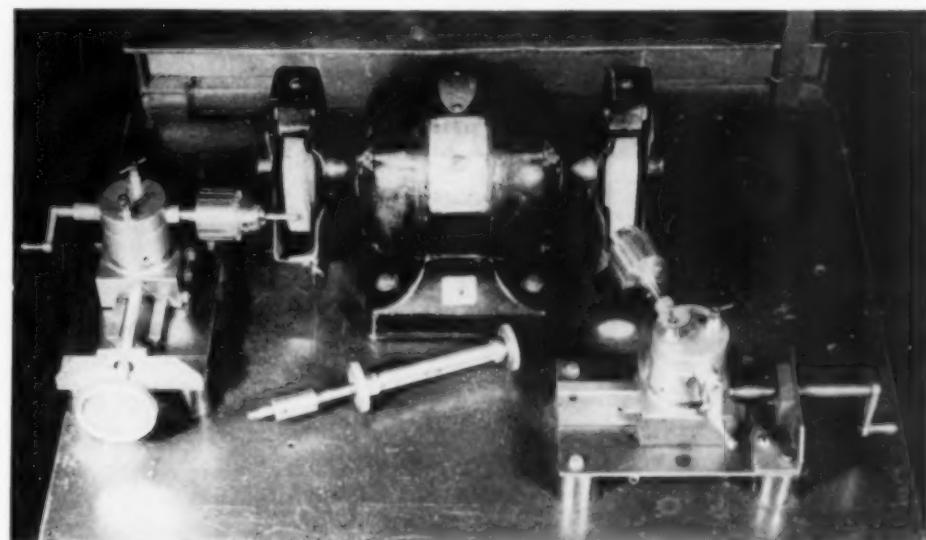


Fig. 2—Protruding Corners of the Chip Are Ground Off on the Left-Hand Wheel, and a Flat Cone Shaped on the Right-Hand Wheel

period of time is multiplied by a factor of five or ten, it is reasonable to expect greater uniformity in the amounts of the various elements which are being vaporized.

The mechanical advantages of this arrangement over a motor-driven interrupter are obvious, and it should be noted that it also acts very effectively as a safety device since it will not allow a voltage to accumulate on the condenser larger than that required to break down the gaps.

From the great volume of experimental work that is going forward in this field of analysis, many improvements in technique and equipment are bound to emerge. These remarks are intended to present some of the solutions that have been found most suitable so far, to meet problems arising in this laboratory, hopeful that they may be of value to others confronted with similar problems.

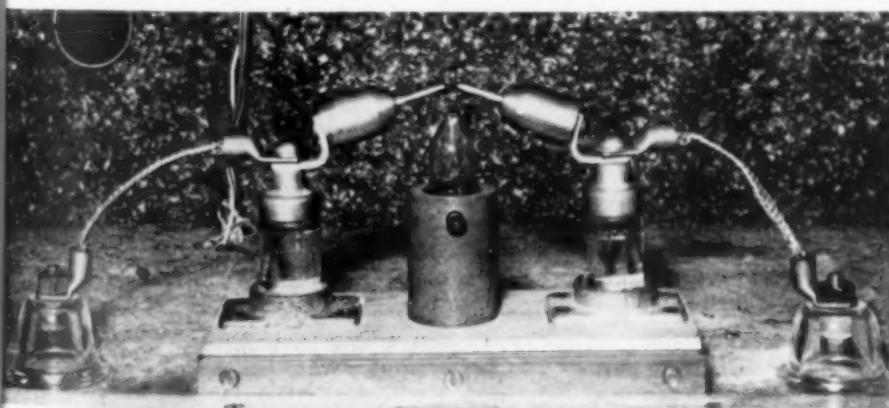


Fig. 3—Air-Gap Interrupter Used in Place of Ordinary Synchronous Interrupter to Increase Largely the Number of Discharges per Cycle and Provide Better Excitation of Flat Surfaces

By J. H. Clark  
J. D. Walker  
and A. S. Jameson

Metallurgists\*  
International Harvester Co.

# NE Steels for Steering Knuckles and Axle Drive Shafts

PRIOR to the critical months of 1941, steering knuckles for farm equipment such as tractors and trucks were being forged from a material whose ranges somewhat resembled the present A.I.S.I. E4132. The main dissimilarity concerned the manganese limits; an openhearth steel with 30 points higher manganese on the maximum side of the range was used for knuckles. This type of material was then replaced by a carbon-molybdenum steel, but by the summer of 1943 it was again changed when, at the completion of a program of testing, it was decided to forge the part from NE9440 steel. It might be added, however, that more recent experimental work suggests the possibility of using NE8632H, with hardenability limitations, sometime in the very near future. (See Table I.)

In the article published in July, reference was made to the basic problem encountered during a replacement program, and it was stated that the initial consideration was one of arriving at a desired hardenability, and from that the required physical properties.

Experimental data accumulated during tests indicated that very little change in heat treatment

\*Of the authors, Mr. Clark is works metallurgist and Mr. Walker is metallurgist at the Fort Wayne Works of International Harvester, and Mr. Jameson is works metallurgist of its West Pullman Works in Chicago. A previous installment appeared in *Metal Progress* last month under the title "NE Steels for Bearings and Bolts in Farm Equipment".

was necessary in order to bring the NE9440 knuckles within the hardness range experienced with the 4132 material. To illustrate this, the knuckles forged from the pre-war 4132 steel were normalized at 1650° F., then heated to 1550° F. with a time quench in water at a controlled temperature of 85 to 90° F. This procedure was followed by a temper that varied according to conditions, but which exceeded a minimum temperature of 975° F. In order to obtain hardnesses within

the range limits of the 4132 steel, it was found only necessary to increase the tempering temperature of the NE9440 to a minimum of 1050° F.

Table I—Average Chemical Compositions

ELEMENT	HIGH MANGANESE E4132	NE9440	8632H
Carbon	0.31	0.41	0.34
Manganese	0.75	1.05	0.75
Silicon	0.28	0.28	0.28
Sulphur (max.)	0.04	0.04	0.04
Phosphorus (max.)	0.04	0.04	0.04
Nickel	....	0.45	0.55
Chromium	0.95	0.40	0.50
Molybdenum	0.20	0.12	0.20
Grain size at 1700° F.	5 to 8	5 to 8	...

In considering their hardenability, it was determined that comparable results could be obtained on both steels. The as-quenched surface and cross-sectional hardnesses of the NE9440 showed Brinell readings that permitted tempering at a temperature sufficiently high to guarantee specified physicals similar to those of 4132 steel. Figure 1 shows the maximum and minimum as-quenched hardness readings on 22 heats of each steel in 1 1/4-in. round bars. It will be seen that the NE9440 shows considerably higher hardenability as quenched. To check this further, surface and cross-sectional hardnesses of

experimental and production-control knuckles of both steels were taken on the heavy end of the spindle at the inner bearing.

The question may arise why a diameter of  $1\frac{3}{4}$  in. was chosen for the hardenability bars represented by Fig. 1. The answer is quite simple. That particular size duplicated the diameter of the heaviest section of the spindle at the inner bearing. This, therefore, presented an accurate picture of the potential hardenability of the material prior to heat treatment of actual parts in production.

Everyone is well aware of the important function steering knuckles perform day after day under varying speeds and road conditions. Consequently, it requires little imagination to realize that a failure of so vital a part could very readily result in a serious accident. For that reason the physical properties of the knuckles are an important consideration. Additional emphasis is placed on this statement when it is realized that the knuckles are closely controlled through actual operational tests on the proving grounds, as well as by metallurgical means during their manufacture.

It might be well to underline the fact that hardness is related to tensile strength. Considering this fundamental truth, it could be reasonably expected that very similar tensile

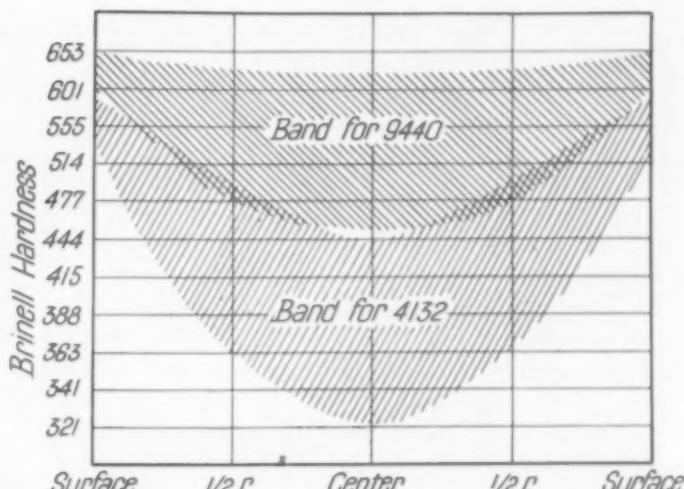


Fig. 1—Maxima and Minima Hardnesses for 22 Heats of Old and Alternate Steels for Steering Knuckles.  $1\frac{3}{4}$ -in. rounds were normalized at  $1650^{\circ}\text{F}.$ , water quenched from  $1550^{\circ}\text{F}.$ , sectioned, and their hardness surveyed

Half of the reported knuckles were forged from 4132 material (9 heats represented) with the remainder being NE9440 steel (2 heats represented). At center and right of Fig. 2 is also shown the relationship of the tensile strength to the yield strength and the reduction of area. As a matter of further comparison, the strength factor for the same two sets of knuckles was computed

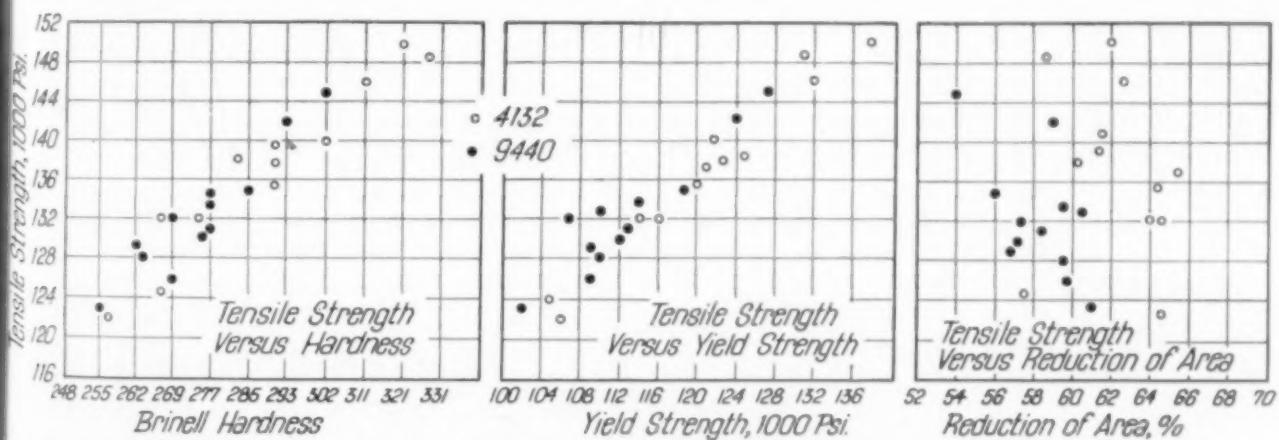


Fig. 2—Relationship of Tensile Strength to Hardness, Yield Strength and Reduction of Area of 0.505-In. Test Bars Machined From Inner Bearing on Completed Steering Knuckles Made of 4132 and 9440 Steels (12 of Each)

properties would be obtained, in view of the hardness results obtained on the two steels.

Figure 2 shows at left the relationship of the core hardness of 24 knuckles, picked at random, to their tensile strength. Hardness was determined by surface grinding a flat on both sides of one end of the 0.505-in. standard test bars, machined from the location of the inner bearing radius. Owing to a taper on the spindle, the bars were taken from the approximate core of the part.

in terms of Schenck's "P" value, where P equals tensile strength divided by 5000, plus  $6/5$  the reduction of area. Results follow:

STEEL:	4132	9440
Maximum value	105.6	99.3
Minimum value	93.9	94.0
Arithmetic mean of 12 tests	101.5	96.2
Standard deviation*	$\pm 2.9$	$\pm 2.0$

\*Limits within which 75% of the observations may be expected to fall.

Tests conducted during the experimental test programs indicated this similarity of physical properties to be a fact, as judged from the response obtained when the two steels were heat treated to the required hardness. Under conditions of adequate production heat treatment it could be concluded that the physical properties of the NE9440 steels were equal to those of the 4132 steels formerly used.

**Proving Ground Tests**—Brief reference has already been made to the fact that performance is additionally controlled by actual test runs on the proving grounds. For the benefit of those unfamiliar with such a test procedure, it might prove beneficial to review briefly the basic principles involved.

When the vehicle has been prepared for a test run through the installation of a test knuckle, an outrigger is attached to the chassis to prevent the unit from turning over. This attachment is adaptable to different chassis and to any model. Figure 3 clearly illustrates the outrigger in position during a test run. The test course is approximately 60 yards long and is built on hard clay ground. Upon completing a test, the course is leveled and the loose ground permitted to settle and harden by rain and sun before another test is conducted.

Tests on knuckles are always operated in conjunction with production material in order to provide a method of comparing the desirability of the test product. A complete test is made with the special steel having been operated on both ends of the axle—right and left hand. If, for instance, the first test is conducted on the

left side, the second will operate on the right. During the run the vehicle is driven in a figure eight. This alternates the maximum side thrust on opposite wheels once each cycle. The figure eight run is reversed every ten laps in order to provide each wheel with maximum thrust at both ends of the course. This is done with the front axle loaded to a rated weight and at a speed of approximately 20 miles per hour before negotiating a curve. At that speed the wheels are sharply cut to the maximum turning angle until the curve is completed. Figure 3 shows such a test vehicle going into a sharp turn.

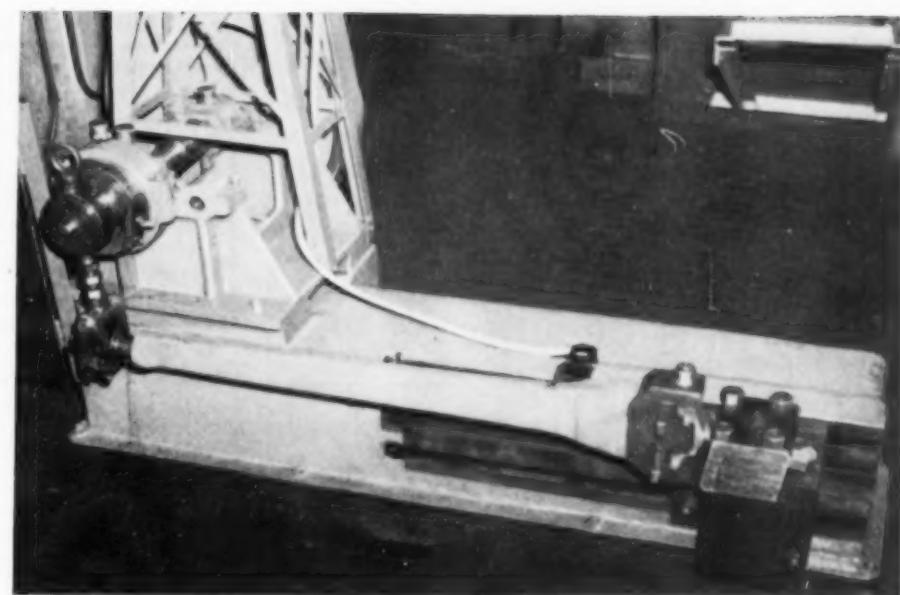
After 30 to 40 miles of this type of driving, the knuckles are removed and inspected for any indication of initial cracking. If no conclusive trend in failure is apparent, the knuckle is replaced and the test run continued until such indications are evident. Production knuckles from former years, of which there is a satisfactory field service background, provide the necessary gage for appraising the new part under consideration. As a result, if the trend of failure during the tests concurs on both new and old axles, we assume that the expected life characteristics of the test knuckles have been achieved.

**Laboratory Endurance Testing**—It is believed to have been said by an ancient Chinese philosopher that a picture was worth ten thousand words. In the interest of economy of space and clarity of description that method of narration has been chosen to describe another type of test on steering knuckles. Figure 4 shows a test stand designed to approximate the bending fatigue loads encountered in service.

*Fig. 3—A Test Vehicle Going Into a Sharp Turn at the Approximate Speed of 20 Miles Per Hour, With the "Outrigger" Preventing the Chassis From Upsetting*



Fig. 4—*Laboratory Stroking Machine Used in Accelerated Tests of Steering Knuckles for Failure in Fatigue. Knuckle end is fixed in jaw at right end of horizontal stroking bar, and spindle end in clamp attached to frame of the testing machine*



Basically, the principle involved is very simple and the fixture is readily adaptable to various size knuckles. It is an accelerated test which causes failure in a comparatively short time by using normal loads, but applying them and releasing them much more frequently than would be encountered in service. When installed in the test position and the test put into operation, the fatigue properties of the knuckles are determined under those alternating axial stresses. After establishing definite cycles to failure, comparisons can be quickly made between standard and alternate steels or designs.

Prior to a testing program of this sort the engineering department determines the conditions of the test and the subsequent load (in pounds) which is applied through the stroking arm by the adjustable eccentric at its left end. Factors involved in the test also control the effective length of the stroking arm in inches. From the known dimensions (including the diameter of the knuckle spindle) it is a matter of arithmetic to arrive at the stress localized at the spindle.

During such a test, each knuckle is stressed to complete failure, with the number of cycles survived being registered automatically. Under standard accelerated test conditions, loads are set to produce failure within 500,000 bending fatigue cycles.

As the result of those test programs, it was determined that the NE9440 steels would give service performances equal to those expected from the pre-war 4132 material. In terms of service life and dependability, that laboratory belief has proven well founded.

#### *Axle Drive Shafts*

In addition to changes in steels for bearings, bolts, and steering knuckles it was necessary to change axle drive shafts from a steel which approximated the latest A.I.S.I. A4145 specifi-

tions, purchased to a special analysis in order to obtain a maximum response to heat treatment. Owing to conditions in the alloy market, it was necessary to substitute A4068 with a restricted manganese range; this was later replaced by NE9450, the steel now being used. The trend of experimental work suggests the adaptability of NE8650H (hardenability specified) for axle shafts.

Everything that has already been written concerning the problem of duplicating the hardenability and resultant physical properties of the steering knuckles with respect to replacement steels can be repeated in the case of the axle drive shafts. Here again the basic consideration was one of arriving at methods of processing in order to meet the physical requirements stipulated for the older standard steel.

As a means of illustrating the similarity of heat treatments given both steels in order to duplicate the hardness ranges, the following information is offered.

Axles forged from both 4145 and NE9450 were normalized at a temperature of 1650° F., then heated to 1525° F. and given an automatically timed pressure quench in oil at 130° F. Both steels were then subjected to an over-all temper, starting at the approximate minimum temperature of 925° F. The tempering temperatures for both steels, naturally, varied with the chemical composition of the heats of steel involved in the processing. As a last operation, the axle flanges were tempered by induction heating to a desired hardness.

Although not based on actual heat analysis but instead on average specification ranges, Table II (on the next page) gives the approximate per-

centage of alloy contained in the old, the present, and the probable future steel.

Using the previously mentioned heat treatment, it was found that the hardenability of the steels in question was quite comparable. Figure 5 represents the as-quenched hardening characteristics of 4145 and 9450 from surface to core. A marked similarity can be seen for the 32 heats of both steels, with the maximum and minimum hardness figures forming the band boundaries. The data were taken from  $1\frac{1}{2}$ -in. round bars.

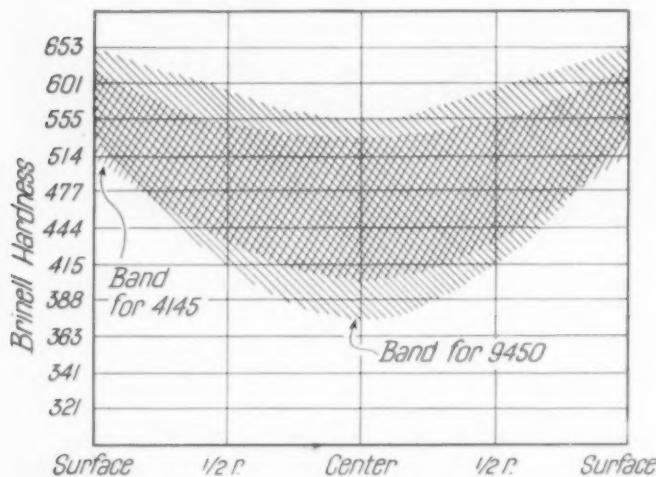


Fig. 5—Maxima and Minima Hardnesses for 32 Heats of Old (4145) and Alternate Steel (9450) for Axle Drive Shafts. Round bars, 1.592 in. diameter, were normalized at  $1650^{\circ}\text{F}$ ., pressure oil quenched at  $130^{\circ}\text{F}$ . from 1550, sectioned, and hardness surveyed

This size approximated the diameter of the axle shafts at the splines, thus permitting a more accurate interpretation of the hardening characteristics of the shafts after production heat treatment.

It is a well known fact that rear axles in

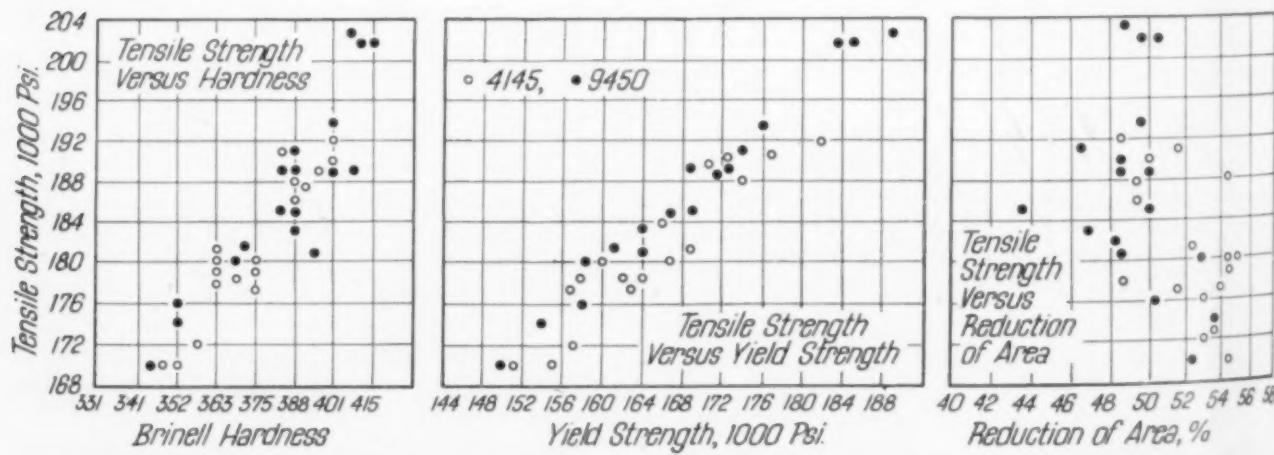


Fig. 6—Relation of Tensile Strength to Hardness, Yield Strength and Reduction of Area of 0.505-In.

Test Bars Machined From the Spline End of Axle Drive Shafts of 4145 and 9450 Steels (18 of Each)

Table II—Steels for Axle Drive Shafts

ELEMENT	S.A.E. 4145	NE9450	NE8650H
Carbon	0.45	0.51	0.50
Manganese	0.75	1.35	0.88
Silicon	...	0.28	0.28
Sulphur	<0.04	<0.04	<0.04
Phosphorus	<0.04	<0.04	<0.04
Nickel	...	0.45	0.55
Chromium	0.95 (a)	0.40	0.50
Molybdenum	0.23	0.12	0.20

(a) Chromium also purchased to special range which permits an average of 1.10%.

vehicles performing heavy duty service are subjected to considerable abuse. This may take the form of severe and repeated overloading, or exposure to operational and service conditions that develop a series of torque or stress applications at or above the yield strength of the material. As a result, it is not at all surprising that such axles occasionally fail.

With this in mind, one realizes the importance of having some means of predicting fairly closely the expected endurance limit of an axle processed from a test material, as well as from steel in regular production.

Certainly some indication of performance characteristics can be drawn from the tensile properties of test bars machined from the finished and heat treated axles and pulled in tension. In examining such figures for strength, it must be kept in mind that the basic criterion of a load-carrying ability is expressed in terms of *yield strength*—that is, the unit load that can be carried without appreciable permanent deformation. Bearing in mind the limitations and drawbacks of a tensile test, due to the fact that the numerical results are based on a standard specimen of 2-in. gage length and 0.505-in. diameter,

and the assumption of a uniform loading over the entire cross-section, additional testing methods were regarded to be necessary for accurate predictions as to the life of an axle. In other words, the physical property values obtained on a small section (test bar) are only true for the large section when the large section itself hardens clear through to the same extent that the test piece does. This is primarily true for the deep hardening steels. It was therefore helpful to incorporate a torsion machine and dynamometer in axle testing programs. This is applied to production control work as well as to preliminary tests for the qualification of new steels.

It has been mentioned that the response of the two axle steels to heat treatment resulted in a similarity of hardness after tempering. Consequently, it would be reasonable to expect that their physical properties would be equally comparable. An interpretation of the results obtained from production and development tests indicated this to be a fact. On the basis of those various

Table III—Torsion Tests on 34 Axle Drive Shafts

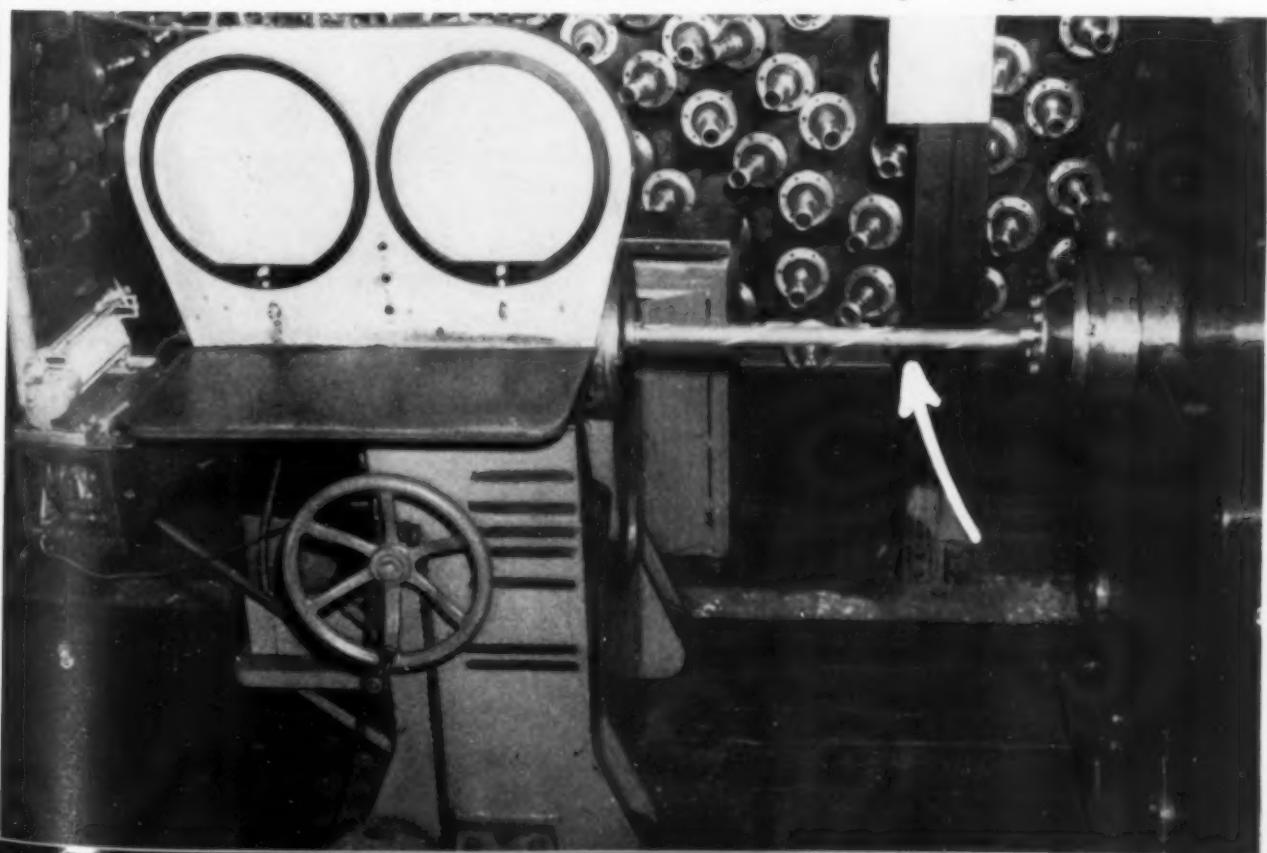
S.A.E. 4145		NE9450	
TWIST, DEGREES	TORSION, IN-LB.	TWIST, DEGREES	TORSION, IN-LB.
1426	110,400	1353	113,800
1466	110,000	1166	110,400
1376	107,600	1364	115,800
1462	111,000	1392	111,900
1524	112,000	1583	115,500
1324	112,000	1431	113,800
1384	115,000	1076	116,300
1611	111,000	491	111,900
1472	112,000	1325	112,300
1486	113,000	535	110,100
1406	119,200	1413	110,200
1593	117,500	1403	104,000
1422	119,800	1474	116,200
1390	121,400	1547	109,400
383	116,800	1183	120,000
1440	117,600	1131	120,000
1353	113,800	1080	120,500
Average	Average	Average	Average
1383	114,125	1232	113,600

tests and subsequent reports from the field, it must be concluded that the physical properties of the NE9450 steels are very similar to those of the original 4145.

**Tension Tests**—By way of illustration, attention is called to Fig. 6. At left is shown the relationship of core hardness of 36 axles to their tensile strengths; 18 represented axles forged from NE9450 (9 heats) with the remainder being processed from 4145 (9 heats). Tensiles and hardnesses were determined from standard 0.505-in. test bars machined from the spline end of the axles. Figure 6 gives additional relationships of the yield and reduction of area to the tensile strength.

**Torsion Tests**—Brief mention has already been made that equipment for axle torsion tests as well as chassis dynamometer equipment enters into the production control and development picture to a great extent. Inasmuch as that equipment has proven to be of considerable importance in the evaluation of test materials, it might be of

Fig. 7—Condition of Axle at the End of a Torsion Test. Arrow points to spiral line, originally a straight mark along the side of the shaft



interest to include a few comments on this phase of the investigation.

Torsion, as the name implies, means the twisting of a body by a couple which turns one end about a longitudinal axis, while the other end is held stationary. Figure 7, page 285, shows an axle torsional machine of 600,000 in-lb. capacity at the end of a test.

To start the test, a specified pre-load is applied and the permanent set of the axle recorded. (This is considered as part of an inspection program, and a percentage of the shafts are periodically checked.) During this phase of the torsion test, the set must not exceed a maximum number of degrees if the inspection specifications are to be met. In applying the pre-load, the diameter of the test axle is taken into consideration. When the permanent set has been recorded, the load is re-applied and the shaft twisted to destruction.

Generally speaking, two axles representing each heat of steel are twisted to destruction as an additional control during production. The maximum load required to induce fracture is recorded in in-lb., as well as the total permanent set or twist. Figure 7 illustrates the result of such a test; the spiral paint line indicated by the arrow on the photograph clearly illustrates the degree of twist.

This test evaluates the suitability of a material for uses which involve lateral as well as longitudinal stresses, where the part must have sufficient torsional strength to support normal

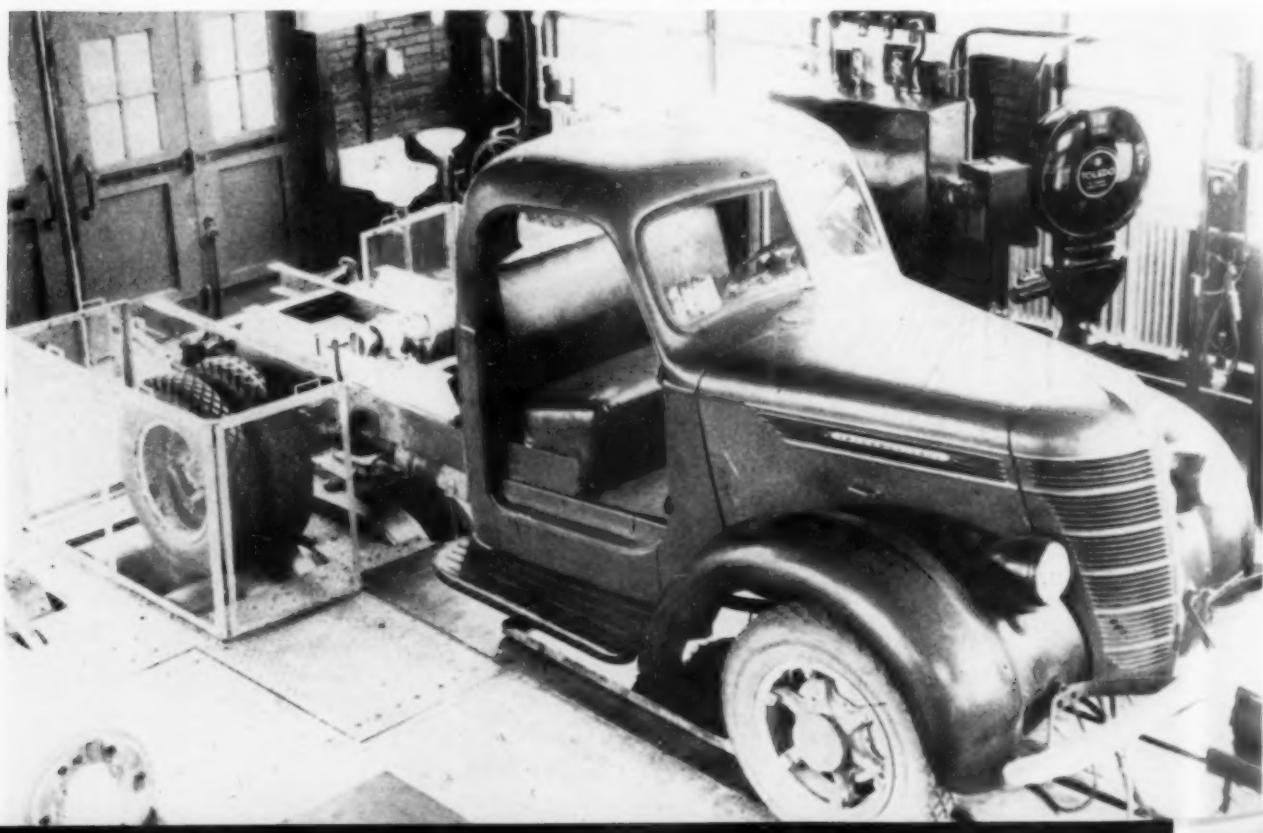
loads as well as occasional moderate overloads.

Table III gives some of the results, and it will be seen that the NE9450 material was comparable to the original axle steel. The figures tabulated involved a total of 34 shafts.

**Dynamometer Tests**—It has already been stated that additional data on the endurance limit of the axles is gathered from the results of chassis dynamometer tests. This testing is done on equipment as pictured in Fig. 8, wherein the test chassis has been driven into position for stationary operation, with rear wheels carefully centered on the rolls. In the background are instrument control panels where pertinent data concerning such variables as engine speed, load, return of load to zero, differential action, tire slippage, and any unusual conditions of chassis or operations are recorded. If, during any investigation that involves axle shafts, transmission gears, differential gears, or any other operational part, failure occurs to a part not under consideration, that part is also examined and data about it gathered for future use.

In subjecting experimental and production control axles to a series of shock loads or clutch impacts, load applications are applied by a sharp disengagement of the clutch. During this procedure the wheels must be completely stopped before each impact. Complying with a designated test pattern, the impacts are transmitted to the axles at a ratio of three forward applications to one in reverse. Naturally, the rate of the impacts vary according to chassis size and differential

*Fig. 8—Test Chassis Mounted on Dynamometer Rollers, Prior to Starting a Test Run. Control panels and recording instruments at right rear*



ratios. However, the average will be between 18 and 22 impacts per min.

As the result of this standardized routine, some comparative figures may be presented. Two groups of tests on axles of 4145 steel were as follows:

	GROUP I	GROUP II
Number of tests	36	25
Average clutching impacts	615	665

Three groups of tests on axles of NE9450 are available showing number of impacts to fracture:

	GROUP I	GROUP II	GROUP III
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Number of tests	35	3	36
Average clutching impacts	475	565	655

A note about the above figures is in order: Since the original data were collected on the 4145 axles, the motor installed in the test chassis has been changed. This resulted in a 5% increase in the maximum torque developed by the motor. As a result, a direct comparison cannot be drawn between the impact resisting values of the NE9450 axles and the 4145 axles; however, it was felt that the average impacts survived were of sufficient interest to warrant including them.

As auxiliary information the "P" values were computed for the axles tested as in Fig. 6—17 from 4145 and 17 from 9450. It will be agreed that these figures indicate an NE steel quite comparable to the older S.A.E. steel with higher alloy:

	S.A.E. 4145	NE9450
Maximum value	106.1	107.6
Minimum value	86.4	89.0
Arithmetic mean	100.0	96.4
Standard deviation	±4.4	±3.9

A thorough analysis of the over-all figures obtained from all the tests, including the dynamometer tests on axles, definitely indicate that the tensile, toughness and endurance limits of the NE9450 are equal to those of the 4145. That such experimental and production data present an accurate interpretation of the fatigue characteristics of the axles is being substantiated day after day under service conditions.

### Transmission Gears

Experimental and field data have indicated that much success has been achieved in adopting the National Emergency steels of the 9400 and 8600 series for transmission gears. It might be said that the majority of the gears in the transmission units now being assembled are being forged from the NE steels of those series. However, in some transmission gears which carry extremely heavy loads, high nickel steels have not been replaced as yet.

As in other critical parts, transmission gears are subjected to load tests designed to impose maximum conditions on the teeth. This type of testing is done in an ordinary dynamometer, as outlined below.

When certain experimental or production control test gears are to be run, they are assembled in a transmission unit and installed on one end of the dynamometer stand, ordinarily with a standard set on the other end. A predetermined load, expressed in terms of ft-lb., is applied. This load represents a certain percentage of the maximum torque of the motor, and as would be expected is governed by the size of the unit under test. Two transmission units are ordinarily operated simultaneously on the stand. They are operated against each other — one clockwise and the other counter-clockwise — with the test gears being installed in the unit running in a clockwise manner. This setup permits the test transmission to be operated at various speeds without the necessity of a variable speed gear box to complete the circuit.

At various intervals, after the break-in period, the general condition of the transmission is examined, with particular attention being paid to the teeth bearings or contacts. As a general rule, the test runs are discontinued when it is quite evident that pitting is prevalent.

Additional data are accumulated on transmissions by installing them in a test chassis, then operating them under shock load or full throttle on the chassis dynamometer. This full throttle test differs from the shock load (impact) test since the gear teeth experience no momentary shock or impact blows. However, they are subjected to combined wear and bending fatigue. From the information recorded on test and production control gears, much valuable data are gathered on the fatigue resistance and mechanical properties of the transmission units.

In conclusion, data have been given in this and the preceding article relating to the substitution of NE steels for those previously used for roller bearings, bolts, steering knuckles, axle shafts, and transmission gears.

The object of all the test procedures was to obtain step-by-step information which would allow the use of the new steels, economical in alloys, during the emergency when alloying metals were scarce, and at the same time produce parts which would be capable of satisfactory performance under normal service conditions. From field reports, there has been no reason to suppose that where the NE steels have been used they have not given service comparable with the original steels they replaced.

# Bits and Pieces

## Polishing Microspecimens for Production Control

**I**N PRODUCTION CONTROL it is desirable to seek speedy and rapid methods, especially since we have a great variety of alloys to examine. The method given below has been developed after trial of various procedures and will be of interest to those who must make routine investigations for inclusion content. For best results, specimens must be hardened in order to retain the inclusions. With the following procedure the specimen can be prepared in 5 to 7 min.

**Rough Polish**—If the specimen is small it should be mounted in some plastic. The specimen is first given a smooth flat grind on No. 1 emery paper on a belt grinder. This should be done slowly to prevent the specimen from heating. Belt grinding is followed by hand polishing on 1/0, 3/0 and worn-out 3/0 metallographic paper, turning the specimen 90° after each paper.

**First Wheel**—The first wheel to be used immediately after the papers is one that is covered with a good grade broadcloth from a tailor shop or dry goods store, securely stretched and fastened. The solution consists of eight teaspoons (heaping) levigated alumina with 30 cc. of liquid soap in 3000 cc. of distilled water. (No. 1557, from Adolph Buehler, Chicago, is quite satisfactory.) This should be thoroughly mixed and decanted in a small flask when needed, leaving behind the very coarse particles. Recommended wheel speed is 700 r.p.m. The specimen should be given a fast circular motion with no side sway. Without circular motion, the inclusions tend to drag or pull out, causing comet tails. About four or five turns with heavy pressure can be applied to the specimen. Then proceed with about four or five turns of light pressure. Do not leave any specimen on any wheel any great length

*"Bits and Pieces" is a department for practicing metallurgists and metallographers, and they are invited to send in short notes about practices that have been found desirable. An  $\odot$  book of your choice is offered for publishable items.*

of time because of the danger of pitting. If scratches are found, it is advisable to go back to the papers and start anew.

**Final Wheel**—Final polishing is done on "Gamal" rubberized polishing cloth using "Gamal" polishing solution with a wheel speed of 250 r.p.m. The best concentration of solution is found by experimenting, and usually is 1 cc. of "Gamal" to 20 cc. of distilled water. (The solution recommended by Fisher Scientific Co. is more dilute and wastes polishing time.) The specimen at this stage is given a radial motion utilizing a side sway motion with medium pressure for a few seconds. Follow this with a few turns of circular motion with slight pressure. One application of polishing solution should be sufficient. A relief polish indicates that the solution is slightly acid; to overcome this, a small amount of a mild alkali such as tri-ethanolamine is added. (C. J. SUMMERFIELD, Metallographer, Babcock & Wilcox Tube Co.)

## Alignment Chart for Computing Diamond Pyramid Hardness Number

**T**HE LIGHT LOADS, precision, and wide range of the diamond pyramid hardness tester have made this tool an indispensable part of our metallurgical research laboratory. The introduction of the Eberbach hardness tester, where loads as low as 7.5 g. may be used, extended this already valuable test into the microscopic region. One of its chief disadvantages, however, is the time required to compute the hardness number from the diagonal length of the impression. Tables are provided for the higher load ranges, but the low load range of the Eberbach must be used with the formula:

$$D.P.H. = \frac{1.8544L}{D^2}$$

where D.P.H. = diamond pyramid hardness number

L = load in kilograms

D = diagonal length, millimeters

This equation can become rather annoying, especially where a large number of readings must be made. The nomogram shown in the data sheet, page 296-B, was designed to solve this equation and has saved considerable time. It is especially

useful where several readings are made with the same combination of load and microscope objective. An additional factor (M scale) was added to allow for the different micrometer calibrations for various eyepiece and objective combinations. The formula then becomes:

$$D.P.H. = \frac{1.8544 L}{(MD')^2}$$

where  $M$  = the eyepiece micrometer calibration in mm. per micrometer division

$D'$  = length of the diagonal in eyepiece micrometer divisions

The length of the diagonal as seen through the microscope and measured with the eyepiece micrometer is used directly in the nomogram without converting to millimeters. The split  $L$  and  $D$  scales enable one to compute either the low load range of the Eberbach or the high load range of the standard Vickers test on the same nomogram with equal facility and with no loss in accuracy. We have found the chart to be well within the accuracy of the load and diagonal length measurements. (HAROLD ROBINSON, Metallurgist, Republic Aviation Corp.)

### Comparative Value of Desiccants

SINCE many dehydrating agents are used (or recommended) for drying furnace atmospheres and other gases used occasionally by metallurgists, a summary of tests made at the National Bureau of Standards may serve as a basis of comparison. Methods of test were

#### Comparative Efficiencies of Various Dehydrating Agents

MATERIAL	VOLUME OF AIR PER HOUR PER MILLILITER OF DESICCANT	TOTAL VOLUME OF AIR PER MILLILITER OF DESICCANT	RESIDUAL WATER PER LITER OF AIR
$\text{CuSO}_4$ (anhydrous)	36 to 50	0.45 to 0.7	2.8 (2.7 to 2.9)
$\text{CaCl}_2$ (granular)	66 to 165	6.1 to 24.2	1.5 (1.4 to 1.6)
$\text{CaCl}_2$ (technical, anhydrous)	115 to 150	4.0 to 5.8	1.25 (1.23 to 1.27)
$\text{ZnCl}_2$ (sticks)	120 to 335	0.8 to 2.1	0.98 (0.94 to 1.02)
$\text{Ba}(\text{ClO}_4)_2$ (anhydrous)	26 to 36	2.3 to 3.7	0.82 (0.76 to 0.88)
$\text{NaOH}$ (sticks)	75 to 170	2.3 to 8.9	0.80 (0.78 to 0.83)
$\text{CaCl}_2$ (anhydrous)	75 to 240	1.2 to 7.8	0.36 (0.33 to 0.38)
$\text{Mg}(\text{ClO}_4)_2 \cdot 3\text{H}_2\text{O}$	65 to 160	4.0 to 7.2	0.031 (0.028 to 0.033)
$\text{KOH}$ (sticks)	55 to 65	3.2 to 7.2	0.014 (0.010 to 0.017)
Silica gel	43 to 59	2.1 to 5.2	0.006 (0.002 to 0.01)
$\text{CaSO}_4$ (anhydrous)	75 to 150	1.2 to 18.5	0.005 (0.004 to 0.006)
$\text{CaO}$	60 to 90	7.6 to 10.1	0.003 (0.003 to 0.004)
$\text{Mg}(\text{ClO}_4)_2$ (anhydrous)	43 to 53	2.8 to 5.9	0.002 (0.0016 to 0.0024)
$\text{Al}_2\text{O}_3$	36 to 63	5.6 to 6.2	0.001 (0.0008 to 0.0012)
$\text{BaO}$	64 to 66	10.6 to 25	0.00065 (0.0006 to 0.0008)

described in the Journal of Research in 1934 (Research Paper No. 649) and need not be repeated here. The results do not always agree with values in published data, which may be due to the fact that drying efficiency varies with different lots of the same agent, depending on details of manufacture and subsequent treatment. Our results are given in the table below. (JOHN H. BOWER, National Bureau of Standards)

### Distinguishing Aluminum Bronze From Manganese Bronze

L. B. CORBETT of the U. S. Bureau of Mines has published an acid test to distinguish between aluminum bronze and manganese bronze. The procedure is as follows: (a) Clean by grinding; (b) spray the spot with dilute sulphuric acid and allow it to react several seconds; (c) add a drop of indicator solution (ammonium-mercury thiocyanate, silver nitrate, ammonium persulphate) and a grayish-purple spot shows manganese bronze, greenish-yellow identifies aluminum bronze. Complete details are given in Bureau of Mines' Report of Investigations, No. 3786.

Anyone having access to a direct-current welding machine can differentiate between these alloys in less time than it takes to grind the surface. Using a carbon electrode, with negative polarity, strike an arc to the unknown sample and *listen!* The arc to aluminum bronze is practically noiseless and the current is steady; the arc to manganese bronze is accompanied by a hissing or crackling sound, due to volatilization of zinc; the intensity of the sound is about the same for the various manganese bronzes and ordinary yellow brass rod. Tell-tale white smoke ( $\text{ZnO}$ ) is evolved from the alloys containing zinc and the current fluctuates. Yellow brass can be distinguished from manganese bronze by its lower hardness.

On Navy G bronzes, the arc makes a noise intermediate between that from manganese and aluminum bronzes, although the color difference and lesser hardness should prevent anyone

from confusing the tin bronzes with these harder materials.

We use the arc method as a routine test to distinguish between aluminum bronze and zinc-bearing silicon bronze (which is a wartime substitute); for some reason the arc from the latter alloy makes the most noise even though some of the other alloys contain three times as much zinc. The spot made by an arc of short duration is so small that the test is non-destructive in many applications. An alternating current arc is too noisy, and even the direct current arc sputters too much on both alloys if the polarity is reversed (electrode positive). (R. W. PARSONS, Chief Metallurgist, The Ohio Brass Co.)

### Tinning a Cast Iron Bearing Back

NOTES on precautions for tinning a steel bearing back or shell were given in *Metal Progress* for July (page 89). The preparation of cast iron boxes represents a special problem, and a chemical bond with the babbitt could not be made — to my belief — until the so-called "Kolene" process was introduced. This requires two baths of proprietary chemicals. One bath is a molten salt bath which runs between 760 and 800° F. This is a very highly oxidizing bath and puts a coating of oxide on the surface. It actually burns out some graphite flakes on the surface of the cast iron, and also has a tendency to decarburize steely parts of the microstructure. As stated before, the less carbon in the iron surface that is to alloy with the tin, the better.

The shell as it comes out of this bath has a dark or black tone. It is then rinsed and placed in the second highly alkaline bath, which is a water solution, that dissolves off this oxide. The shell is again rinsed and given a very quick dip in a cold 10% hydrochloric acid which neutralizes any remaining alkali. Without further rinsing, the shell is then placed into zinc-chloride liquid flux, and then into the tinning bath.

The salt used for the first bath is quite hygroscopic. Also, the alkali bath evolves ammonia in the reaction and therefore must be placed under a hood. The results obtained are so good that we have been willing to put up with the difficulties of handling the baths. For example, we make a number of crosshead shoes for large gas engines and these have always had grooves machined in them to hold the babbitt. Today we have eliminated grooves completely and have cut the thickness of the babbitt from  $\frac{3}{8}$  to  $\frac{1}{16}$  in. with a great deal of success. (T. E. EAGAN, Chief Metallurgist, Cooper-Bessemer Corp.)

### Simple Equipment for Electropolishing Stainless Steel Microspecimens

SPECIAL TECHNIQUES, such as given by A. L. SCHAEFFLER in the August, 1944, issue of *Metal Progress* (page 285) are required to produce scratch-free surfaces on stainless steel microspecimens. It is well known that electropolishing will produce scratch-free surfaces, but the recommended solutions require either heating or cooling, and are generally inconvenient for the general metallurgical laboratory.

When attempting to etch a stainless steel specimen electrolytically in concentrated nitric acid the writer noticed that the specimen became covered with a reddish-brown film; after the film was removed the surface was lustrous and



*Stainless Steel Dish Is Cathode, Pointed Rod Is Anode for Electrolytic Polishing and Etching Solutions*

scratch-free. Based on this discovery, a rapid method has been set up. (When inclusions are to be studied this method is unsatisfactory, as no method of electropolishing will preserve them.)

Procedure is as follows:

1. Mount in plastic (for small specimens, wires, and so on).
2. Belt grind (80 and 150-grit "Try-M-It" belts).
3. Machine polish (150-grit "Aloxite" and Baker's chrome oxide).
4. Electropolish in conc.  $\text{HNO}_3$  for 30 sec.
5. Electro-etch in 10% oxalic acid for 1 min.

The above illustration is evidence of the simplicity and convenience of the apparatus. The anode contact is made by means of the pointed rod. The stainless steel dish serves as the cathode. (OTTO ZMESKAL, Research Metallurgist, Universal-Cyclops Steel Corp.)

A Critical Review  
by Frank G. Norris  
Asst. Metallurgical Engineer  
Wheeling Steel Corp.  
Steubenville, Ohio

## Basic Slags and Dephosphorization

CERTAIN PROBLEMS of steel manufacture keep recurring from time to time. For a long time steelmakers have tried to describe a "good" slag and to duplicate it from heat to heat and to distinguish it, preferably in some simple way, from a slag that does not quite do the job. The latest information published on this general problem of slag and slag-metal relations is in three advance copies of the *Journal of The Iron and Steel Institute of Great Britain*. Each of these papers is very thorough and will repay careful study of the original, as only the high points can be covered in this review. Their complete titles and authors are as follows:

"The Constitution of Basic Steel Furnace Slags", by J. R. Rait and H. J. Goldschmidt, of the research department of William Jessop and Sons, Ltd., Sheffield.

"A Study of the Basic Open-Hearth Process, With Particular Reference to Slag Constitution", by A. H. Jay, of the central research department of the United Steel Companies, Ltd., Stocksbridge.

"The Phosphorus Reaction in Basic Open-Hearth Practice", by Y. K. Zea, of Steel, Peech and Tozer, Sheffield.

Messrs. Rait and Goldschmidt made an X-ray study of open-hearth and electric furnace slags by the powder method, and to this reviewer it seems a bit disappointing that so much work is done on a slag with the merest mention of the nature of the steel

that was made under it. This is no doubt an example of what happens if a subject is divided into compartments and one assumes that after studying each compartment very minutely he will arrive at a general view of the whole problem. Any major investigation into the chemistry of the basic steel-melting processes must include a study of the constitution of slags and refractory linings and also the slag-metal, slag-refractory, and metal-refractory relationships.

There is always the question of how best to bridge the gap between the liquid slag in the furnace and the solid slag that is studied. Rait and Goldschmidt's investigation is based on the premise that a knowledge of the chemical and mineralogical constitution of the crystallized slags is a primary essential to the study of the slags in the liquid state and to the major study of the chemistry of steelmaking. The further importance of constitution of basic slags is that its rapid determination could be used as the basis of a scientific control of the slag during melting and refining. Another application of slag constitution is of greater importance in British practice than in most parts of America, namely, the use of basic slags for fertilizer because of the recovery of phosphate.

Chemical analysis separates the slag composition into various oxides; the principal ones in basic slags being CaO, FeO, MnO, MgO, SiO<sub>2</sub>, P<sub>2</sub>O<sub>5</sub>, Fe<sub>2</sub>O<sub>3</sub>, and Al<sub>2</sub>O<sub>3</sub>. Some slags, particularly basic electric

oxidizing slags, may contain small amounts of Cr<sub>2</sub>O<sub>3</sub>. "Slag chemistry" is concerned with the relations prevailing within equilibrium systems of these oxides, and "slag constitution" refers to the combinations or compounds of the above mentioned oxides existing at furnace temperatures.

There are several rather widely used methods for identifying crystal substances. These include the examination of thin polished sections by the petrographic microscope, the determination of refractive indices, and the actual separation and chemical analysis of the phases present. In Rait and Goldschmidt's investigation the X-ray powder (Debye-Scherrer) method was used, wherein the constitution can normally be determined readily by matching diffraction patterns against standard photographs. The powdered slag is mounted on a hair with Canada balsam. A particle diameter of 1 1/4 mm. (approx. 14 mesh) is the optimum size and improves the detectability of weak lines.

The powder method yields the following information:

1. All slags examined were largely crystalline, and contained but small amounts of glass (amorphous or non-crystalline materials).

2. Phases were identified and their relative amounts estimated. A quantitative method of estimation was developed for basic electric reducing slags.

3. The composition of the solid solution (comprised largely of FeO, MgO, MnO, and CaO) was estimated from lattice parameter measurements by comparison with a known calibration curve.

4. The range of composition of the phases was estimated. The degree of homogeneity (if there be "degrees" of homogeneity) is shown by the breadth of a spectral line.

Twelve standard compounds were prepared. The method of preparation of the pure oxides is not as simple as may appear at first thought. Though not of detailed interest to the general

reader, research workers may well refer to this section of the paper.

Basic electric reducing slags are described first because they are less complex than oxidizing slags. The complete chemical analysis and the calculated constitution is given for 42 slags arranged in order of  $\text{CaO}/\text{SiO}_2$  ratio from 2.82 ( $\text{CaO}$  62.30%,  $\text{SiO}_2$  22.10%) to 1.32 (42.80 : 32.40). The main components of these slags are  $\text{CaO}$  and  $\text{SiO}_2$  with small amounts of  $\text{MgO}$  and  $\text{Al}_2\text{O}_3$ ; the sum of these four amounts to 90% or more. Iron, manganese, and sulphur are minor components. When spar is absent the compositions of reducing slags fall within the  $\text{CaO}\text{-MgO}\text{-SiO}_2\text{-Al}_2\text{O}_3$  quaternary system. (The various three-component systems are presented and crystallization within each is discussed in detail.) No slags with  $\text{CaO}/\text{SiO}_2$  below 2.00 contain the compound  $3\text{CaO}\cdot\text{SiO}_2$  or  $3\text{CaO}\cdot\text{Al}_2\text{O}_3$ , and most slags below and only one above the 2.00 ratio contain  $\text{MgO}\cdot\text{Fe}_2\text{O}_3$  and  $\text{MgO}\cdot\text{Al}_2\text{O}_3$ .

In certain compositions in the  $\text{CaO}\text{-MgO}\text{-SiO}_2$  ternary the calcium orthosilicate  $2\text{CaO}\cdot\text{SiO}_2$  becomes what is known as a "disappearing phase" because, although it is first formed on cooling, the final composition under equilibrium at

room temperature is merwinite ( $3\text{CaO}\cdot\text{MgO}\cdot2\text{SiO}_2$ ), akermanite ( $2\text{CaO}\cdot\text{MgO}\cdot2\text{SiO}_2$ ) and monticellite ( $\text{CaO}\cdot\text{MgO}\cdot\text{SiO}_2$ ). There are several other illustrations of the phenomenon of the disappearing phase. This type of crystallization path (showing disappearing phases) has an important influence on the resultant constitution of a slag, depending on its rate of cooling. Such a slag can obviously have quite different constitutions with different rates of cooling.

Another application of these phase diagram studies, that is only mentioned in passing, is the study of the powdering or "falling" of certain high-lime blast furnace slags on cooling, which of course makes the slag useless for such products as road ballast.

Since fluorspar is added to a considerable number of the steel-making slags, its influence is discussed briefly by Rait and Goldschmidt. Several slag compositions were mixed with an equal weight of fluorspar and fired two days at  $1200^\circ\text{C}$ . ( $2200^\circ\text{F}$ .) and then examined by X-rays for the phases present. Apparently  $2\text{CaO}\cdot\text{SiO}_2$  is not decomposed by spar, but  $3\text{CaO}\cdot\text{SiO}_2$  loses all of its silica to form free  $\text{CaO}$ . Merwinite

( $3\text{CaO}\cdot\text{MgO}\cdot2\text{SiO}_2$ ) loses part of its silica by being converted to  $2\text{CaO}\cdot\text{SiO}_2\cdot\text{CaF}_2$ , does not form solid solutions with either  $\text{CaO}$  or  $2\text{CaO}\cdot\text{SiO}_2$ , and therefore must form simple eutectic systems with these compounds.

The reducing slags are classified as follows:

**Lime Slags**—Highly basic white slags containing free lime and  $3\text{CaO}\cdot\text{SiO}_2$  as the major compounds, with smaller amounts of  $\text{MgO}$ ,  $\text{CaF}_2$ , and  $3\text{CaO}\cdot\text{Al}_2\text{O}_3$ . These slags are avoided because they tend to be viscous in the furnace and require spar for rapid fluxing and suitable fluidity. They also tend to build up the bottom because of the difficulty in draining the heavy slag. The chief advantage is that increasing basicity of the slag increases the efficiency of sulphur removal.

**Falling Slags**—The normal type of reducing slag disintegrates to a fine powder on cooling due to the transformation from the  $\beta$  to the  $\gamma$  form of the large amounts of  $2\text{CaO}\cdot\text{SiO}_2$  present in these slags. The inversion occurs at  $675^\circ\text{C}$ . ( $1250^\circ\text{F}$ .) with a 10% increase in volume.

**Merwinite Slags**—When there is a delay in tapping, the white falling slag becomes brown and

*Testing Fluidity of Slag, an Important Aid to Good Steelmaking. (Courtesy Bethlehem Steel Co.)*



stable. These slags all contain merwinite ( $3\text{CaO}\cdot\text{MgO}\cdot2\text{SiO}_2$ ) and usually no other compound. Their lime-silica ratio is lower than in falling slags; apparently the silica dripping from the roof during holding at high temperature changes the  $2\text{CaO}\cdot\text{SiO}_2$  of the falling slag to  $3\text{CaO}\cdot\text{MgO}\cdot2\text{SiO}_2$ . The  $\text{CaF}_2$  content of these slags is low, confirming the observation that spar is decomposed in the presence of merwinite. Merwinite slags can of course be formed at an earlier stage if lime additions are insufficient. In general these slags are undesirable because of their greater solubility for  $\text{MgO}$ , which consumes much dolomite refractory; likewise they are poor desulphurizers.

**Carbide Slags**—Calcium carbide  $\text{CaC}_2$  usually persists only in highly basic slags containing free lime. In the slags examined the amount of carbide was too low to be detected by the X-ray method.

**Ladle Slags**—The constitution and properties of the tapping slags have an important influence on the life of ladle linings. The slag reacts with the ladle lining, and its alumina and silica content increases. Gehlenite ( $2\text{CaO}\cdot\text{Al}_2\text{O}_5\cdot\text{SiO}_2$ ) is one of the compounds formed and adheres to the ladle lining, possibly finding its way into subsequent casts of steel.

In summary, Rait and Goldschmidt discuss the various mineral phases in basic electric refining slags and give a brief history of their discovery and the characteristics by which they are recognized. These phases are:

1. Free oxides; (a)  $\text{MgO}$ , below about 1.8  $\text{CaO}/\text{SiO}_2$  some, or in a few cases all, of the  $\text{MgO}$  is combined; (b)  $\text{CaO}$ , found only in a few very basic slags.

2. Tri-calcium silicate, stable only between 1250 and 1900° C. (2280 and 3450° F.) decomposing at other temperatures into  $\text{CaO}$  and  $2\text{CaO}\cdot\text{SiO}_2$ . Its rate of decomposition is so low that it can exist indefinitely at ordinary temperatures. (The statement that an unstable compound can exist indefinitely involves some mental gymnastics. Like other gymnastics, if you are in training you can take them right in your stride, but otherwise you are liable to have a charlie horse the next day!)

3. Calcium orthosilicate ( $2\text{CaO}\cdot\text{SiO}_2$ ).

4. Merwinite ( $3\text{CaO}\cdot\text{MgO}\cdot2\text{SiO}_2$ ) and its analogue, manganese-merwinite ( $3\text{CaO}\cdot\text{MnO}\cdot2\text{SiO}_2$ ).

5. Calcium fluoride, usually associated with  $\text{CaO}/\text{SiO}_2$  over 1.4.

6. Spinel ( $\text{RO}\cdot\text{R}'\text{O}_3$ ). Only one slag contained enough  $\text{MgO}\cdot\text{Al}_2\text{O}_5$  for identification. The divalent metal  $\text{R}''$  may be  $\text{Mg}$ ,  $\text{Fe}$ , or  $\text{Mn}$  and the trivalent  $\text{R}'''$  may be  $\text{Al}$ ,  $\text{Fe}$ ,  $\text{Mn}$ , or  $\text{Cr}$ .

7. Melilite can be present only with rare conditions—a low  $\text{CaO}/\text{SiO}_2$  ratio and all of the  $\text{MgO}$  in combination.

8. Monticellite ( $\text{CaO}\cdot\text{MgO}\cdot\text{SiO}_2$ ) can only occur with  $\text{CaO}/\text{SiO}_2$  less than 1.4 and therefore seldom appears in basic finishing slags. The  $\text{Mg}$  is sometimes replaced by  $\text{Fe}$  or  $\text{Mn}$ .

9. Other phases. Some other phases theoretically possible were not detected, probably because of the rapid cooling; the last liquid is quenched to glass. Thus, the amount of  $\text{CaS}$  and  $\text{CaC}$  were too small to be detected by X-rays.

### Constitution of Basic Electric Oxidizing Slags

The chief components of oxidizing slags for ordinary carbon steels are the basic oxides  $\text{CaO}$ ,  $\text{FeO}$ ,  $\text{MnO}$ , and  $\text{MgO}$ , and the acid oxides  $\text{SiO}_2$ ,  $\text{Fe}_2\text{O}_3$ ,  $\text{Al}_2\text{O}_5$ , and  $\text{P}_2\text{O}_5$ . A system of phase assemblages, assuming complete crystallization, is shown in Rait and Goldschmidt's paper for four conditions, two of them based on the additional presence of spar or  $\text{Cr}_2\text{O}_3$ . The chemical analysis and mineralogical constitution calculated according to the systems of phase assemblages are given for 32 slags.

Calculated and X-ray results are very similar, discrepancies being chiefly due to the fact that the slags—because of their rapid cooling—are not completely crystallized, whereas the calculations are based on complete crystallization under equilibrium conditions. In other words, with rapid cooling some or possibly all of a compound may fail to crystallize, in which case it is dissolved in the glass, and such constituents are not detected by X-rays.

Naturally the phase equilibria have not been worked out for the eight-component system. It is difficult for the casual reader to appreciate either the amount of work involved in determining even a ternary equilibrium dia-

gram or the amount of information that is revealed to those who can interpret one. As an example, in this situation lack of thermal equilibrium data prevents any sound deductions to be made as to the effects of rapid cooling on the constitution of the slag. A simplification is possible by studying the pseudo-ternary system:  $\text{CaO}\cdot2\text{CaO}\cdot\text{SiO}_2\cdot4\text{CaO}\cdot\text{Al}_2\text{O}_5\cdot\text{Fe}_2\text{O}_3$ .

Rait and Goldschmidt classify the slags into five chief types according to the main phases detected by the X-ray method:

Type 1. Highly basic slags in which  $3\text{CaO}\cdot\text{SiO}_2$  and  $\text{RO}$  solid solutions are the main phases. Provided they contain sufficient iron oxide, these slags remove phosphorus most efficiently and because of their low solubility for basic oxides they are least severe on the dolomite banks.

Type 2. Most of the slags have  $\beta\ 2\text{CaO}\cdot\text{SiO}_2$  and  $\text{RO}$  solid solutions as the chief phases. "Falling" slags are absent, probably because of the stabilizing effect of small amounts of  $\text{P}_2\text{O}_5$ . This slag is fluid without spar, removes phosphorus efficiently and does not attack the banks severely.

Type 3. A group of slags with lower  $\text{CaO}/\text{SiO}_2$  ratio than the above. Small amounts of  $3\text{CaO}\cdot\text{RO}\cdot2\text{SiO}_2$  probably stabilize the  $2\text{CaO}\cdot\text{SiO}_2$ . This modified high temperature form is referred to as  $\alpha'\ 2\text{CaO}\cdot\text{SiO}_2$ . The other main phase is the  $\text{RO}$  solid solution.

Type 4. Merwinite ( $3\text{CaO}\cdot\text{MgO}\cdot2\text{SiO}_2$ ) is the outstanding phase accompanied by small amounts of  $\text{RO}$  solid solution and  $2\text{CaO}\cdot\text{SiO}_2$ . The combination of  $\text{RO}$  in merwinite decreases the amount of free  $\text{RO}$  solid solution as the decreasing  $\text{CaO}/\text{SiO}_2$  ratio shifts the slag from Type 3 to Type 4. Merwinite slags are not as efficient phosphorus removers as the more basic slags, and the higher solubility for  $\text{MgO}$  increases the dolomite consumption. These slags should therefore be avoided for normal basic electric melting.

Type 5. Merwinite is again the outstanding phase accompanied by monticellite ( $\text{CaO}\cdot\text{MgO}\cdot\text{SiO}_2$ ). Negligible amounts of phosphorus are removed by this type of slag. It is very fluid and, being unsaturated with respect to  $\text{MgO}$ , is severely corrosive to the dolomite banks. It should be avoided in the oxidation stage of normal basic electric melting.

There are intermediate stages between these groups, but the above is a practical classification. (The phases occurring in basic electric oxidation slags are described in detail in this paper.)

The phase assemblages and methods of calculation are not given here because those sufficiently interested to want to make the calculation for their own slags will doubtless look up the original paper. Others may not be concerned with these details.

### Constitution of Basic Openhearth Slag

Basic electric oxidizing and basic openhearth slags have the same component oxides and therefore belong to the same phase systems. Phosphate in the electric slags is too low for X-ray identification. In contrast,  $P_2O_5$  is one of the major acid radicals in basic openhearth slags and considerable amounts of phosphates occur. (Remember that the above statement applies to British practice; many of their electric oxidizing slags have from 0.50 to 1.50%  $P_2O_5$ , and therefore openhearth slags of American practice using Lake ores are closer to their electric than to the Britishers' openhearth slags.)

Four phase assemblages in solidified openhearth slags are given for four conditions:  $SiO_2/P_2O_5$  ratio over 180:142, from 180:142 to 60:142, less than 60:142 and those containing fluorine, assuming the exact amount of fluorine for combining with all of the  $P_2O_5$  as apatite to make "fluorapatite" ( $9CaO \cdot CaF_2 \cdot 3P_2O_5$ ).

Partial chemical analysis is given for 43 slags; all but six were supplied by three steelworks. The slags from each works are arranged in increasing order of  $CaO/FeO$  ratio as determined by the X-ray patterns and starting with slags with no free lime.

The phase constitution cannot be calculated from the partial analyses. The essential validity of the system of phase assemblages is shown by calculation of the constitution of 12 slags studied by B. Mason in his article on "The Constitution of Some Basic Openhearth Slags" in the *Journal of the Iron and Steel Institute* for 1944, II, p. 65 p, and the agreement between observed and calculated results is striking. Free

lime is the most prominent phase in most basic slags.

For efficient production the constitution of the slag should be controlled within fairly close limits. When the lime content is too low, removal of phosphorus and sulphur is inadequate and the slag cuts the bottom by dissolving  $CaO$  and  $MgO$ . When the lime content is too high spar is added, thus destroying the availability of phosphate as a fertilizer.

At the lower lime contents all of the lime is in combination as silicates and silico-phosphates. With higher basicity free  $CaO$  appears as a separate phase, the amount increasing with increasing basicity. The ferrous oxide is uncombined and, with  $MgO$  and  $MnO$  in solution, forms a separate

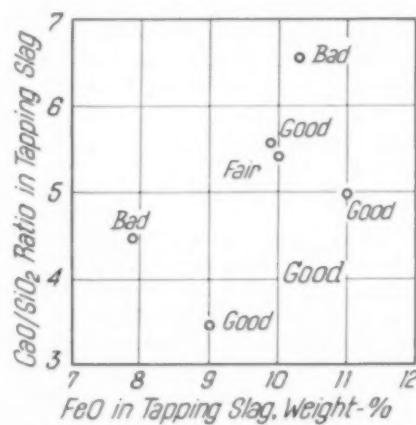


Fig. 1—Relation Between  $CaO/SiO_2$  Ratio and FeO Content in Tapping Slags

phase. Thus a satisfactory normal basic openhearth slag should show free  $CaO$ , and  $(Fe, Mn, Mg)O$  in addition to the silicates and silico-phosphates.

These two free oxide phases can be readily identified and the amounts approximately estimated by the X-ray powder method. Thus by determining the free  $CaO$  and  $RO$  phases in a number of slags by the X-ray powder method and correlating these results with the steelmaking practice, the ideal range of free  $CaO$  and  $RO$  can be established, and additions of lime and ore may be made as required based on X-ray examination during the process. The X-ray powder method would form a perfectly sound method of control, provided the results could be obtained rapidly. This rather important proviso is not discussed in great detail, though several technical

shortcuts are suggested. The recently developed techniques of self-reading optical spectrographs may eventually be adapted to X-ray analysis.

Let us summarize briefly: X-ray studies have identified the several phases present in solid slags. The amount of these phases can be calculated from complete slag analysis and certain knowledge of co-existing phases and cooling conditions.

Slag analysis can be reported in two ways, either as the usual chemical analysis, or as percentages of various compounds. Now the practical question arises: Does the knowledge of the percentage of mineral compounds justify the time and trouble of getting this information?

We already know that a "lean skinny slag" will cut the banks. Scientifically it is quite satisfying to know that this action is due to merwinite, but does this knowledge in any way affect the corrective action—which would be to add more lime? Some of you may answer, "Yes," or others, "This clue may lead to improve-

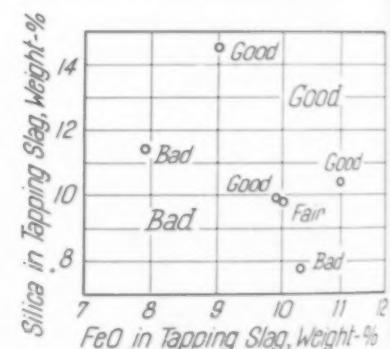


Fig. 2—Relation Between  $SiO_2$  and FeO Contents in Tapping Slags

ment in practice if we follow it just a bit further." Still others may answer, at least mentally, "Furnace operation is already sufficiently complicated without having to wonder about different minerals that may not even exist in the liquid slag." In other words, the individual's answer is a matter of opinion that is affected by all of the contributions of outlook and individual manner of thinking that influence differences of opinion on any subject.

The paper by A. H. Jay of the United Steel Companies central research department refers both to basic openhearth slag constitution and the dephosphorization

process. In regard to dephosphorization Mr. Jay dismisses the variations between the fundamental assumptions of Schenck\* and the conclusions based on X-ray studies with the statement that each paper (his own and Mr. Zea's, abstract of which follows) is a contribution to our knowledge on a problem which still warrants considerable study. This abstractor will not undertake to untangle the complications which baffle the author.

In Jay's paper both slag and metal samples are included for several stages of the heat. Because of the effect of cooling rate the mineralogical constitution of both water quenched and spoon cooled samples is given. Several supplementary experiments were made because the presence of the usual phases did not fully account for the X-ray line pattern. Phases were obtained from this auxiliary study which fully account for all of the lines on the X-ray photographs of basic slags.

To determine the constitution of molten furnace slag it was decided to quench a small amount

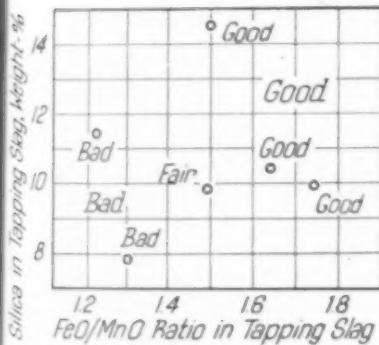


Fig. 3—Relation Between  $\text{SiO}_2$  Content and  $\text{FeO}/\text{MnO}$  Ratio in Tapping Slags

of slag in water. Such slag was quenched not more than 3 sec. after withdrawal from the furnace. The lime-rich phase is strikingly well crystallized and it is therefore thought to be present as a solid component in molten furnace slag. The other phase in quenched slags was the lime-silica phosphate phase called Nagelschmidtite ( $9\text{CaO} \cdot \text{P}_2\text{O}_5 \cdot 3\text{SiO}_2 \cdot 7\text{CaO} \cdot \text{P}_2\text{O}_5 \cdot 2\text{SiO}_2$ ). No other pattern was observed. The results are interpreted as showing that the solubility of  $\text{MnO}$  and possibly  $\text{FeO}$  and  $\text{MgO}$  is higher at

\*H. Schenck and W. Riess, Archiv für das Eisenhüttenwesen, 1936, p. 589.

steelmaking temperatures than at lower temperatures.

Another crystalline phase of the  $\text{FeO} \cdot \text{MgO} \cdot \text{MnO}$  type appeared when quenching was slower. Its poor crystallinity suggested that it was not present as a separate phase in the molten slag. This evidence seems to be in direct contradiction to the report of Rait and Goldschmidt of the co-existence of  $(\text{CaMn})\text{O}$  and  $(\text{Fe}, \text{Mn}, \text{Mg})\text{O}$  in basic slags ranging from no free lime to high lime content. The difference suggests, not that either investigation is wrong, but that knowledge of the quenching conditions is required for interpreting the X-ray films.

The detailed study of slags from several heats may be summarized as follows: At the clear-melted stage the lime is low, increasing as refining proceeds; the silica is high and decreases.

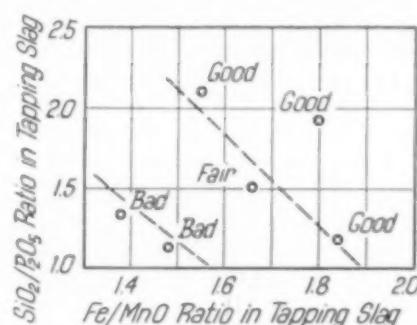


Fig. 4—Relation Between  $\text{SiO}_2/\text{P}_2\text{O}_5$  Ratio and Total  $\text{Fe}/\text{MnO}$  Ratio in Tapping Slags From Steels Containing 0.45 to 0.65% of Carbon and 0.7 to 0.85% of Manganese

The iron content of the clear melted sample showed marked variations; these variations are associated with the residual phosphorus existing at the time—being of the order of 0.04% P, which is considered low (the pig iron contained about 5% P.)  $\text{MnO}$ ,  $\text{P}_2\text{O}_5$ , and  $\text{MgO}$  usually decrease as refining proceeds, owing to dilution by the increasing slag weight. Magnesia may increase when a bottom boil is present.

### Rephosphorization

Figures 1 to 6 from Dr. Jay's paper illustrate the development of a hypothesis expressing the probability of rephosphorization of the pit samples. All tapping slags contained free lime.

The X-ray findings indicate that silica combines with lime to

form a phase of the  $2\text{CaO} \cdot \text{SiO}_2$  type in preference to combining with  $\text{FeO}$  to form fayalite ( $2\text{FeO} \cdot \text{SiO}_2$ ) or with  $\text{MnO}$  to form either tephroite ( $2\text{MnO} \cdot \text{SiO}_2$ ) or rhodonite ( $\text{MnO} \cdot \text{SiO}_2$ ). Thus the ratio  $\text{CaO}/\text{SiO}_2$  is used as one factor in the first chart. The presence of a phase rich in ferrous oxide in the spoon cooled slag led Jay to try "FeO in the tapping slag" as the second factor (see Fig. 1). In this a fair separation of the good from the bad heats is noted, but the unsatisfactory position of one good sample shows that other factors should be considered. Because the slag samples had a nearly constant lime content, Fig. 2 was drawn using silica as the vertical ordinate instead of the  $\text{CaO}/\text{SiO}_2$  ratio used in Fig. 1.

In both Fig. 1 and 2 the possible effect of  $\text{MnO}$  is not considered, yet there is reason to believe that it may retard the oxidizing power of  $\text{FeO}$ . Therefore in Fig. 3 the oxidizing power (horizontal ordinate) is represented by the  $\text{Fe}/\text{MnO}$  ratio. This results in an improved separation of good from bad heats.

Thus far the  $\text{P}_2\text{O}_5$  content has been ignored, though it obviously has an important bearing on rephosphorization. Therefore in Fig. 4  $\text{SiO}_2/\text{P}_2\text{O}_5$  is plotted against the  $\text{Fe}/\text{MnO}$  ratio. Note the use of *total iron* rather than  $\text{FeO}$ , the change being for no good reason except that it is stated to be more in keeping with normal laboratory methods of analyzing slag samples. Figures 5 and 6 are the extension of the method of Fig. 4 to two groups of heats, one containing about 0.5% C and 0.75% Mn, and the other 0.25% C and 1.05% Mn.

The relations shown by these diagrams can be expressed thus:

1. For a given  $\text{Fe}/\text{MnO}$  ratio rephosphorization is retarded by increasing the  $\text{SiO}_2/\text{P}_2\text{O}_5$  ratio.

2. For a given  $\text{SiO}_2/\text{P}_2\text{O}_5$  ratio rephosphorization is retarded by increasing the  $\text{Fe}/\text{MnO}$  ratio. In other words, a high oxidizing slag must be maintained.

These indications require that at least a part of the  $\text{P}_2\text{O}_5$  should be held as a solid lime-silica phosphate in the molten slag. Thus it is inferred that molten basic slag is a two-phase system containing a solid and a liquid, not unlike a suspension.

The process of slag formation is outlined by equations indicat-

ing that the fluxing or solution of lime sets free  $FeO$  and  $MnO$  from their silicates. In other words, in the presence of excess lime, ferrous and manganous silicates are not stable and cease to exist.

Further study indicated that other factors such as the turbulence during casting are involved. To study this effect an experiment was devised in which the metal and slag were represented by two immiscible liquids such as oil and water, water and carbon tetrachloride, and water and bromoform; these substances during a

high magnesia content acts to inhibit the oxidizing power rather than to increase the basicity. (As none of the slags run over 7.5%  $MgO$ , the expression "unusually high" magnesia is slightly ambiguous. Whether it means slags considerably over those reported or simply the higher of the reported slags is not clear. In any event this statement, while not proving that it is wrong to add together  $CaO + MgO$  in the appraisal of a given slag, should serve as a caution before adopting this simplification, used by some workers in plotting and studying basic slag compositions.)

We are hearing more and more about the influence of the ladle lining, as was mentioned briefly in the above paper. Now comes the third paper by Y. K. Zea on the phosphorus reaction in the basic openhearth, in which the flat statement is made; "Rephosphorization does not occur when the ladle is lined with basic brick." The degree of rephosphorization with a fireclay brick lining may be controlled to a large extent by keeping the temperature as low as possible and by increasing the

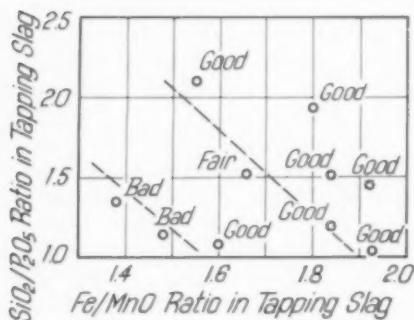


Fig. 5—Relation Between  $SiO_2/P_2O_5$  Ratio and Total  $Fe/MnO$  Ratio in Tapping Slags From Ten Heats of Steel Containing 0.45 to 0.65% of Carbon and 0.7 to 0.85% of Manganese

violent agitation and mixing give opportunity for action between "slag" and "metal".

The reversion of phosphorus during the time the metal leaves the furnace and is cast into the first ingot should be related to the amount of liquid phase. Reversion during pouring the heat should be controlled by the stability of the lime-silica phosphate phase. Rise of phosphorus content of the last ingots is presumably the result of reaction with the firebrick of the ladle.

If the slag has too high lime, its oxidizing power decreases. This is shown by high residual phosphorus and reversion, and by the small content of the  $FeO$ -type phase in the spoon cooled slag. Therefore a well balanced slag has an optimum range of free lime and either too much or too little contributes to high phosphorus.

Increase of magnesia can cause rephosphorization in the furnace. Magnesia reacts with the liquid phase and increases the viscosity of the slag. Also the  $MgO$  associates itself with  $FeO$  rather than with lime. Thus, an unusually

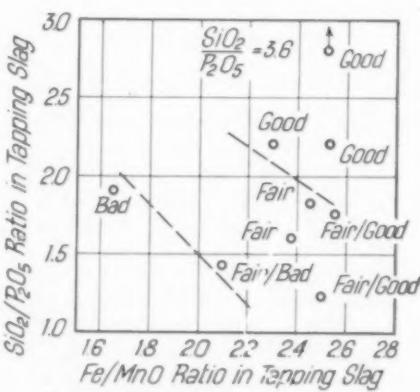


Fig. 6—Relation Between  $SiO_2/P_2O_5$  Ratio and Total  $Fe/MnO$  Ratio in Tapping Slags From Steels Containing 0.25 to 0.30% of Carbon and 1.05 to 1.10% of Manganese

lime content so that even after reaction with the ladle the basicity of the slag will not fall too low for it to hold the phosphorus.

This, in brief, is a summary of the results of Dr. Zea's paper. The rest of the paper is difficult to abstract; the original should be studied to do it justice.

It has been believed for a long time that slag analysis can be used to compare various slags and to study their effect upon the metal.

Then came the idea of assuming (usually without very rigorous proof) that the silica and phosphorus are combined with lime in simple compounds. Schenck seems to have progressed furthest in this direction. He expresses some of the characteristics of a slag in terms of free  $CaO$  and free  $FeO$ . The effect of temperature is included in these concentration terms—that is, temperature changes the amount of free  $CaO$  by affecting the dissociation of the compounds of the slag. Thus far this concept is quite simple, though in practice considerable skill and patience are required to use Schenck's diagrams. The chief justification for this idea and for the assumed combinations is that the system seems to work, so that predicted reactions are confirmed closely by analysis. Failure of X-ray studies to identify some of the compounds is a bit disconcerting to those who believe that evidence from all sources should fit into one systematic pattern.

Dr. Zea has selected 15 heats to determine the soundness of various theories of dephosphorization by testing the accuracy in practical application. There is little doubt that, at least for these 15 heats, the average difference and the standard deviation of the difference between the calculated and the analyzed phosphorus are lower when the phosphorus is calculated by the method of Schenck and Riess.

Finally, it is worthy of remark that Zea's investigation is another made possible by developments in high temperature thermometers for measuring the bath temperature. A quick immersion pyrometer was used. The author mentions that "the importance of accurate temperature measurement can hardly be over-emphasized, since lack of a precise temperature record leads to misinterpretation and miscalculation".

As can be realized by considering the number of theories that are advanced, the last word has not yet been said on the constituents of basic slag and their effect on reaction with the metal. The present position is well summarized by one of the authors: "There is no direct means as yet available to determine the true constitution of molten basic slags..."

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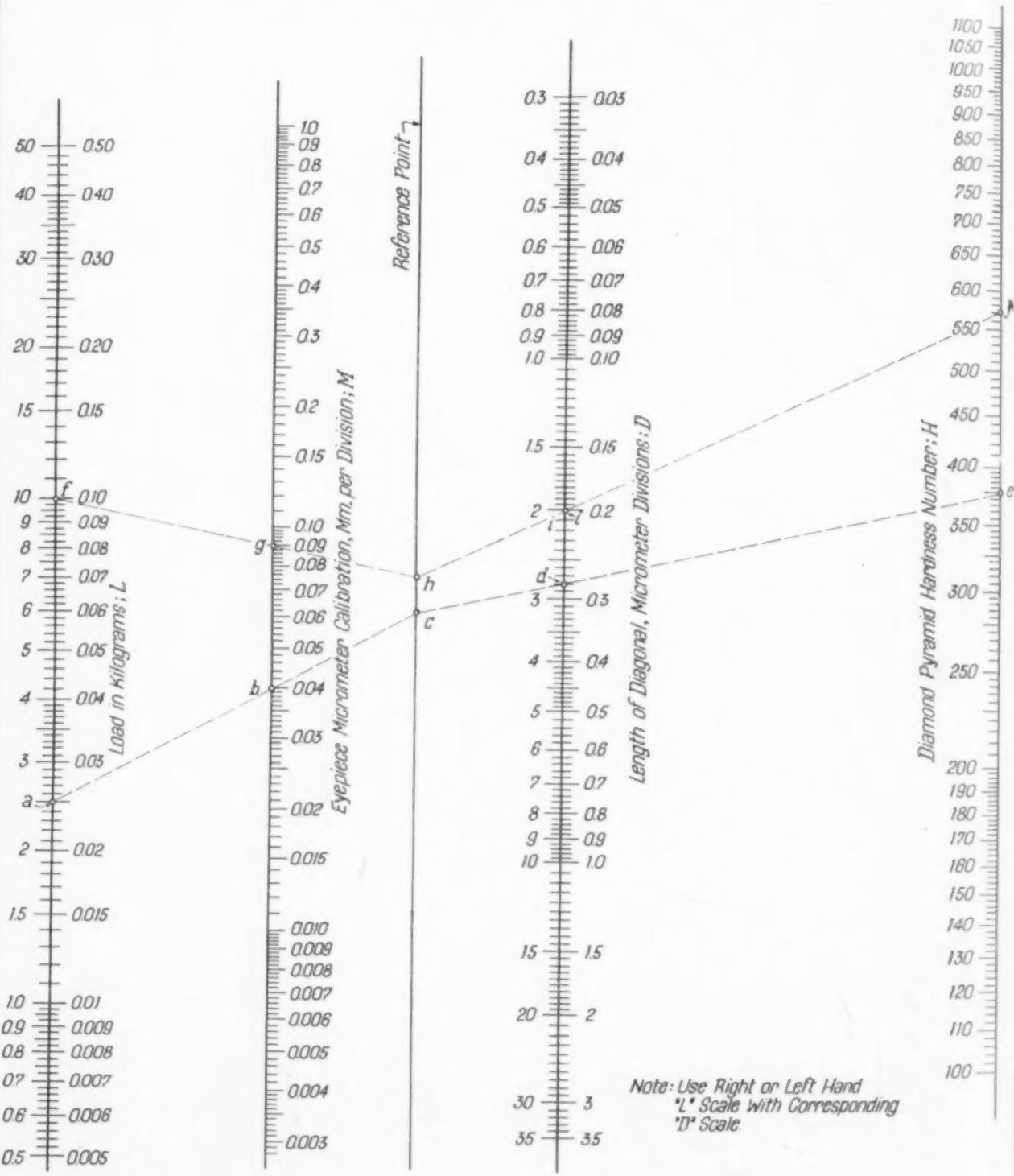
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# Nomogram for Diamond Pyramid Hardness

(Vickers or Eberbach)

By Harold Robinson



## Eberbach Tester

Connect load *a* (25 g.) with eyepiece micrometer calibration *b* (0.04) and project line to point *c*. Connect *c* with observed length of diagonal impression *d* (0.28 division on eyepiece micrometer) and extend to scale *H* for hardness number: 375 (point *e*).

## REPRESENTATIVE SOLUTIONS

## Vickers Tester

Connect load *f* (10 kg.) with eyepiece micrometer calibration *g* (0.09) and project line to point *h*. Connect *h* with observed length of diagonal impression *i* (2.0 divisions on eyepiece micrometer) and extend to scale *H* for hardness number: 570.

By Louis A. Carapella\*

*Mellon Institute,  
Pittsburgh*

# Fundamental Alloying Nature of Magnesium

DURING the past two decades an increasing propensity toward the use of scientific principles and away from the ancient arts has been especially noteworthy in the field of metallurgy. Metallurgists of today, with some knowledge of equilibrium diagrams of alloy systems, can predict certain relationships between the properties and the alloy diagrams. Prior to this advancement in physical metallurgy very little information was available on the factors governing the type of equilibrium diagrams that would result whenever two or more metals were alloyed. Through recent extensive and systematic consideration of reliable experimental data, certain useful and fundamental principles have been expounded concerning the behavior of alloy systems.

Inasmuch as magnesium has gained a prominent position as an industrial metal during the past few years, its future potentialities appear to be tremendous as long as research continues to look for new and better alloys. It is therefore the prime objective of this article to stimulate more interest in this direction by reviewing and elaborating upon certain fundamental principles of alloying behaviors of magnesium. In this manner, it would then be possible to ascertain fully the nature and deportment of magnesium in combination with other metals, and thereby to acquire a working knowledge by which certain physical properties can be predicted for a given alloy combination, or by which new magnesium

alloys can be fabricated with certain desirable physical properties.

Let us proceed, therefore, to state some of these fundamental principles of alloying:

Wherever magnesium alloys with other metals, it forms what is described as a "substitutional solid solution". Such a solid solution is one in which atoms of the solute metals replace at random atoms on the magnesium lattice. The extent to which this replacement may take place is termed "solid solubility". The several factors which have been found to govern the limit of solubility in alloy systems will now be discussed with special reference to magnesium and its binary alloys.

**Type of Crystal Structure**—Magnesium is a close-packed hexagonal metal, and therefore complete solid solubility—that is, ranging from pure magnesium, through all percentages of the second metal—can only be expected with other hexagonal metals, provided all other factors are favorable. Metals with other crystal structures cannot form complete solid solutions with magnesium, for by their very nature they introduce at least one phase to the system that is not hexagonal. Nevertheless, non-hexagonal metals may exhibit considerable solubility in magnesium for reasons which will be considered later.

**Relative Atomic Sizes**—Where the atomic sizes of magnesium and solute metals differ by less than 15%, the size factor is favorable. If the atomic sizes exceed this limit, the solubility is restricted. In fact, the greater the difference in size, the more restricted is the solubility, if other factors are equal. Solute atoms which have atomic sizes just on the edge of the favorable zone tend to give erratic results. Moreover, if the atomic sizes differ by more than 8% but are still within the favorable zone, there is usually a minimum in the liquidus curves, representing a

\*Senior Fellow on the magnesium development program sustained in the Institute by the National Lead Co. of New York.

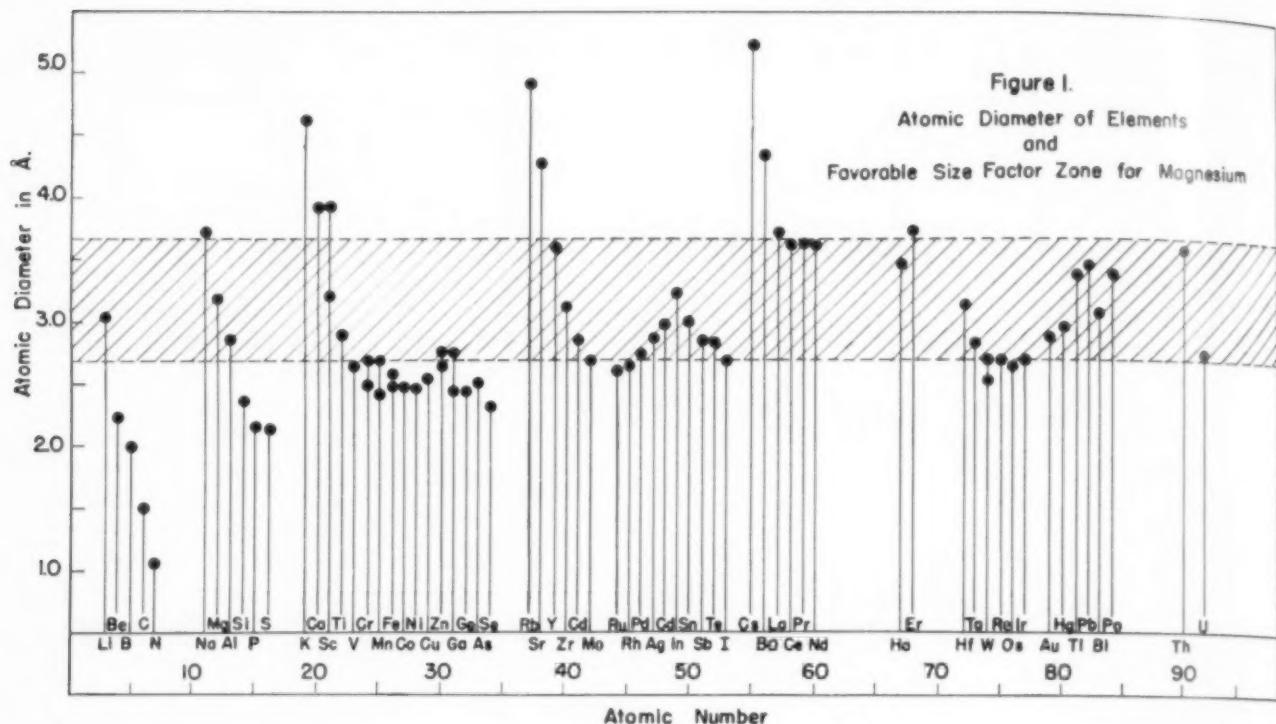


Figure 1.  
Atomic Diameter of Elements  
and  
Favorable Size Factor Zone for Magnesium

definite tendency toward a eutectic formation.

In instances where the size factor restricts the solubility of one metal in magnesium, there is usually a tendency for solubility to increase with rising temperature, so that the system may show age-hardening characteristics. Therefore the size factor may facilitate the choice of combinations of metals likely to strengthen magnesium alloys through age hardening.

**Valence (Electron) Factor** — Magnesium is a divalent solvent and, where the size factor is favorable, tends to form a wide range of solid solutions with other divalent metals because of equal valency. There is a general trend (where the size factor is favorable) for magnesium solid solutions to become more restricted as the valencies become more unequal. Furthermore, the more unequal the valency factor, the steeper is the drop in both the liquidus and solidus curves.

The solidus curves are much more affected than the liquidus curves of magnesium binary systems wherever there is an increase in either the valence or the atomic size of alloying metals, or both. It is therefore possible by the proper selection of alloying metals to reduce wide freezing ranges — which are generally undesirable in foundry alloys.

**Electrochemical Factor** — The more electro-positive the solvent metal and the more electro-negative the solute metal, or vice versa, the greater is the tendency to restrict solid solubility and to form intermetallic compounds. The electronegative degree of metals in the periodic sys-

tem of chemical elements increases from left to right in any period and from bottom to top in any group.

Magnesium is more highly electropositive in nature than most industrial metals. This characteristic results in the formation of intermetallic compounds with most metals, and these compounds are formed at the expense of solid solutions. The tendency for formation and the stability of the compounds increases, the more electronegative the alloying metal. Thus, even if the size factor is favorable, the solubility of a metal in magnesium may be restricted by the formation of a compound. The extent to which this may occur depends essentially on the stability of the intermetallic compound. Since the general tendency of solid solutions is to an increase in solubility with rising temperature, alloys with a propensity toward the formation of compounds may be expected to manifest age-hardening properties, for the solubility will be less at the lower temperatures and a hardening precipitate will form.

### Discussion of Basic Principles

Even though the size factor is very important, it is obvious from the foregoing brief statements that it alone does not sufficiently establish whether or not a metal will alloy with magnesium, but it does serve as a kind of preliminary test which must be satisfied before a solid solution can be formed. Thus it will enable us to

rule out certain metals as being unlikely to form solid solutions; attention can then be given to the more probable alloys.

**Atomic Size Factor** — The application of the rule about size factor may be better understood from Fig. 1, in which the atomic diameters of different elements are plotted against their respective atomic numbers. As the atomic "diameters" of elements are not fixed values, but since the space an atom occupies depends largely on the nature of interatomic forces within the crystal lattice containing it, each atom may have

ties of magnesium are somewhat circumscribed.

**Modified Periodic System of Elements** — As a collective representation of the fundamental factors affecting the alloying nature of magnesium, a modified periodic chart is set forth as Table I. The size factors of elements with respect to magnesium are indicated either as favorable, border line, or unfavorable, and are accompanied by their respective magnitudes in percentage.

Because the crystal structure of elements determines the extent of solid solubility to some degree, it is also included in the table. Where

Table I — Periodic System of Elements —  
Their Structures, Alloying Valences, and Atomic Size Factors With Reference to Magnesium

	I		II		III		IV		V		VI		VII		VIII		
	a	b	a	b	a	b	a	b	a	b	a	b	a	b	a	b	
I																	
2	3 Li ○(-5%) BCC		4 Be ●(-30%) HCP				5 B ●(-38%) Hex?		6 C ●(-53%) D-C*		7 N —		8 O —		9 F —		2 He —
3	11 Na ○(+6%) BCC		12 Mg ○(0%) HCP				13 Al ○(-10%) FCC		14 Si ●(-26%) D-C		15 P ●(-32%) FCO*		16 S ●(-34%) FCO		17 Cl —		10 Ne —
4	19 K ○(+45%) BCC		20 Ca ●(-23%) FCC*		21 Sc ●(-23%) FCC*		22 Ti ○(-9%) HCP*		23 V ●(-18%) BCC		24 Cr ○(-15%) BCC*		25 Mn ○(-15%) BCC*		26 Fe ●(-19%) BCC*	27 Cr ●(-22%) HCP*	28 Ni ●(-22%) FCC*
	29 Cu ●(-20%) FCC		30 Zn ○(-13%) HCP		31 Ga ○(-3%) FCO		32 Ge ●(-23%) D-C		33 As ●(-21%) Rho.		34 Se ●(-27%) Hex*		35 Br —			36 Kr —	
5	37 Rb ●(-5%) BCC		38 Sr ●(-35%) FCC		39 Y ○(+14%) HCP		40 Zr ○(-1%) HCP*		41 Nb ○(-7%) BCC		42 Mo ○(-15%) BCC		43 Ta —		44 Ru ●(-7%) HCP	45 Rh ●(-16%) FCC	46 Pd ○(-14%) FCC
	47 Ag ○(-10%) FCC		48 Cd ○(-7%) HCP		49 In ○(-6%) FCT		50 Sn ○(-12%) BCT*		51 Sb ○(-10%) Rho.		52 Te ○(-10%) Hex.		53 I —			54 Xe —	
6	55 Cs ○(+60%) BCC		56 Ba ●(-36%) BCC		57-71 Rare Earth Metals		72 Hf ○(-6%) HCP		73 Ta ○(-11%) BCC		74 W ○(-15%) BCC		75 Re ○(-15%) HCP		76 Os ●(-16%) HCP	77 Ir ○(-15%) FCC	78 Pt ○(-13%) FCC
	79 Au ○(-10%) FCC		80 Hg ○(-6%) Rho.		81 Ti ○(-10%) HCP*		82 Pb ○(-9%) FCC		83 Bi ○(-3%) Rho.		84 Po ○(-7%) M'cl?		85 Ab —			86 Rn —	
7	87 VI —		88 Ra —		89 Ac —		90 Th ○(+2%) FCC		91 Pa —		92 U —						
Alloying Valence	I	I	2	2	0	3	0	4	0	5	0	6	0	—	0	0	—

Code:-  
 ○ Favorable Size Factor  
 ○ Border line " "  
 ● Unfavorable " "

BCC- Body Centered Cubic  
 FCC- Face " "  
 D-C- Diamond Cubic  
 HCP- Hexagonal  
 BCT- Body Centered Tetragonal  
 Rho- Rhombohedral  
 M'cl- Monoclinic

\* Other Structures also Exist.

different volumes under different circumstances. In the figure, wherever the atomic sizes differ greatly the two limits are indicated; otherwise only the most probable value is presented.

The favorable atomic size factor zone with reference to magnesium is shown by the shaded area in Fig. 1. Of all the elements under consideration, about half of them are within the favorable zone, about one-tenth are on the border line, and the remainder are outside the favorable limits. On this basis alone, the alloying possibil-

more than one modification is known, the structure which ordinarily exists is listed, and an indication that others are known is made wherever possible.

The general valency behavior of elements during alloying is appended to each group. Normally metals have valences equal to the number of free electrons they possess. Transition metals, however, owing to their characteristic incomplete inner electronic structures, act as though they have zero valency wherever alloyed

with "normal" metals, in this sense.

Finally, the electrochemical effect may be estimated from the relative positions of metals in the periodic system. There is a tendency for metals to form normal valence compounds with the metals of Group IV-b to VI-b, inclusive. As the metals from these groups become more electronegative with respect to magnesium, their tendency to form compounds with higher stabilities is greater.

### Experimental Evidence

In the preceding sections of this article the principles concerning the general nature of alloy systems have been listed. A brief consideration of experimental facts in the light of these principles will now be applied to alloy systems of magnesium that have been studied. The amount of reliable information available is limited, but nevertheless it is sufficient to ascertain something about the alloying nature of mag-

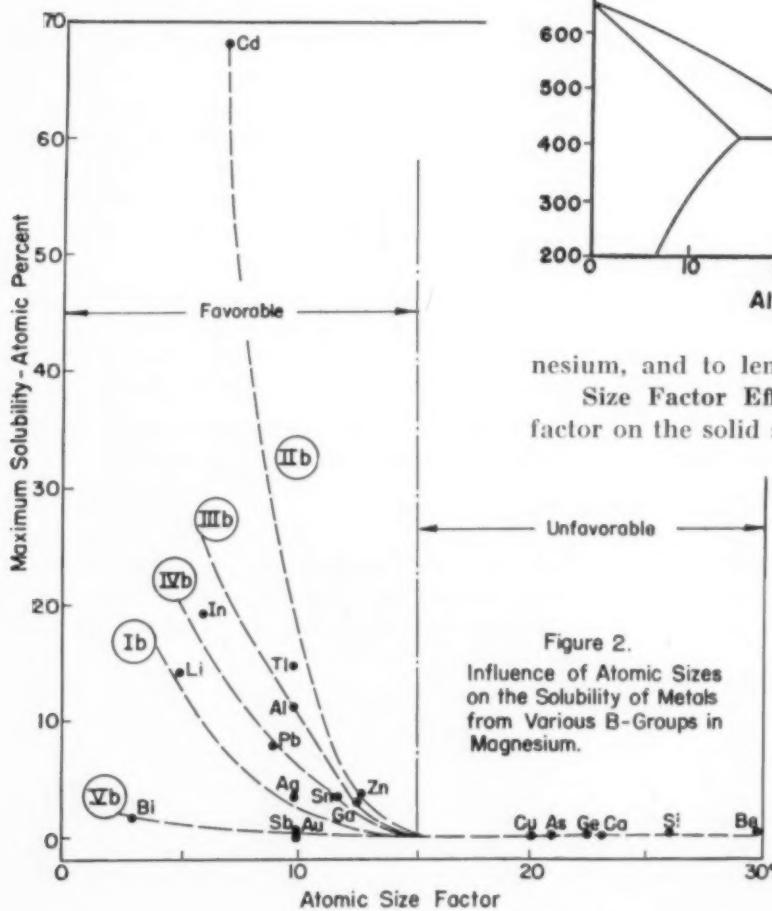


Figure 2.  
Influence of Atomic Sizes  
on the Solubility of Metals  
from Various B-Groups in  
Magnesium.

Figure 3.

### Influence of Size Factor on Liquidus, Solidus, and Solid Solubility Curves

#### Supplementary Data

##### A. Aluminum

1. Crystal Structure FCC
2. Size factor -10% (favorable)
3. Alloying Valence 3
4. Group III b; period 3

##### B. Gallium

1. Crystal Structure FCC
2. Size factor -13% (favorable)
3. Alloying Valence 3
4. Group III b; period 4

##### C. Indium

1. Crystal Structure FCT
2. Size factor -6% (favorable)
3. Alloying Valence 3
4. Group III b; period 5

##### D. Thallium

1. Crystal Structure HCP
2. Size factor -10% (favorable)
3. Alloying Valence 3
4. Group III b; period 6

\* See Table I.

nesium, and to lend some authority to tentative predictions.

**Size Factor Effects** — The importance of the atomic size factor on the solid solubility of metals in magnesium is apparent from Fig. 2. Within the favorable zone, the solubility of metals from any one group increases as the size factor decreases, or as the atomic sizes approach that of magnesium. It should also be remembered that the change of solubility with atomic size factor is different for each group but is in accordance with the relative valence effect.

For a given atomic size, metals in Group II-b are more soluble in magnesium than metals from other groups. Moreover, it is interesting to note that in general the maximum solubility of a

Group I-b (univalent) metal is less than that of a Group III-b (trivalent) metal; and yet the difference in valence with respect to magnesium is the same for metals in both groups. (This effect may be explained in terms of the newly proposed Brillouin zone, or electronic band theory of metals.) As the atomic size factor exceeds 15%, the solubility becomes quite restricted and is independent of valence effect.

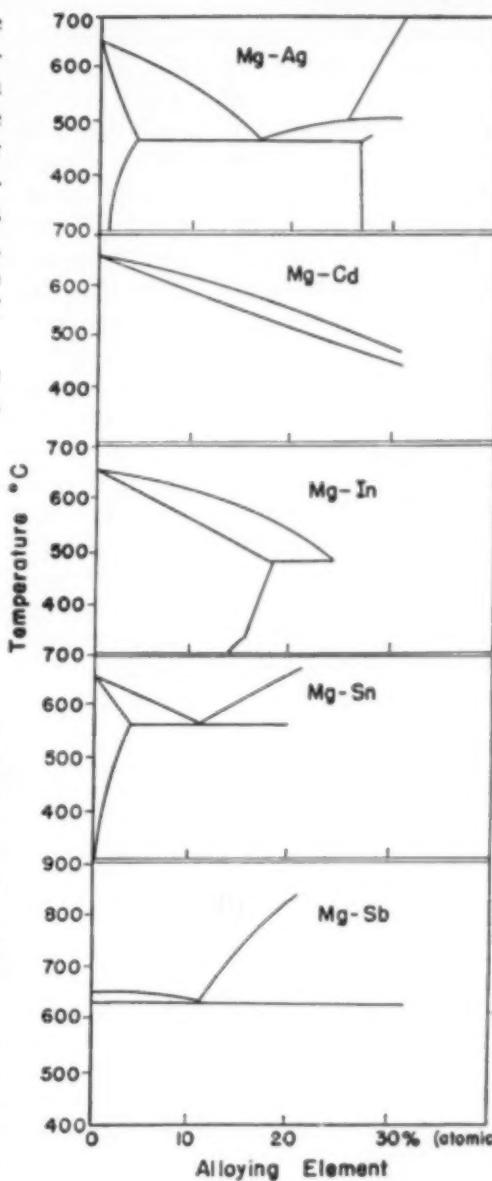
The influence of atomic size factor on the liquidus and solidus lines from binary alloys of magnesium with metals in Group III-b is illustrated in Fig. 3. As the four alloying metals in this figure are from the same group, the valence factor is restricted, and so the effects shown in the diagrams may be attributed chiefly to the atomic size factor. Both the liquidus and the solidus curves from the magnesium-rich portion of the systems are greatly depressed as the atomic size factors differ increasingly, the solidus curves being generally more affected than the liquidus curves. Wide freezing ranges may therefore be expected wherever the atomic size factors become unfavorable.

**Valence Factor Effect** — Similar influences on the liquidus and solidus curves of alloy systems are experienced through the action of the valence factor. In Fig. 4 the effect of this factor may be seen as demonstrated by alloy systems of magnesium with fifth period metals from various groups. Undoubtedly a more conclusive demonstration of the influence of this factor alone could be made if it were possible to select alloy systems of metals from the same period with equal atomic sizes. This condition cannot be fully realized because of the inherent nature of the atoms (see Fig. 1), although the general effect is apparent. Both the liquidus and solidus curves are greatly affected as the difference in valences between magnesium and the alloying metal increases, and the solidus curves are again more susceptible to the action of the valence factor than the liquidus curves.

**Electrochemical Effects** — Apart from the valence effect that is normally experienced by

Figure 4.

Influence of Alloying Valence on Liquidus, Solidus, and Solid Solubility Curves



Supplementary Data

A. Silver

1. Crystal Structure FCC
2. Size factor -10% (favorable)
3. Alloying Valence 1
4. Group Ib; period 5

B. Cadmium

1. Crystal Structure HCP
2. Size factor -7% (favorable)
3. Alloying Valence 2
4. Group IIb; period 5

C. Indium

1. Crystal Structure FCT
2. Size factor -6% (favorable)
3. Alloying Valence 3
4. Group IIIb; period 5

D. Tin

1. Crystal Structure BCT
2. Size factor -12% (favorable)
3. Alloying Valence 4
4. Group IVb; period 5

E. Antimony

1. Crystal Structure Rhbdi.
2. Size factor -10% (favorable)
3. Alloying Valence 5
4. Group Vb; period 5

alloying metals in different groups with magnesium, another factor is introduced which exerts a strong influence on the behavior of alloy diagrams depending on the relative electrochemical nature of the metals. Reference is had to the electrochemical factor, the importance of which may be seen in the alloy systems of magnesium with Group IV-b metals illustrated in Fig. 5 on the next page. This factor promotes compound formation. If this tendency is great, compounds will form regardless of whether the atomic size factors are favorable or unfavorable. The greater the influence of this factor, the more stable is the resulting compound. (Stability is normally indicated by the melting point of the compound.)

**Thermodynamic Aspects** — From a thermodynamical consideration of the free energy of

phases likely to exist in eutectiferous alloy systems, one can explain the influence of the electrochemical factor on the nature of equilibrium systems on the basis of compound stability. The two phases shown in each of the diagrams in Fig. 5 may be in equilibrium with each other at definite compositions for a given temperature only if the total free energy is a minimum with respect to any internal changes of the system. A common tangent is drawn to the free energy curves, and this determines the compositions of both phases which can co-exist in equilibrium, because the total free energy of the system is a minimum under these conditions.

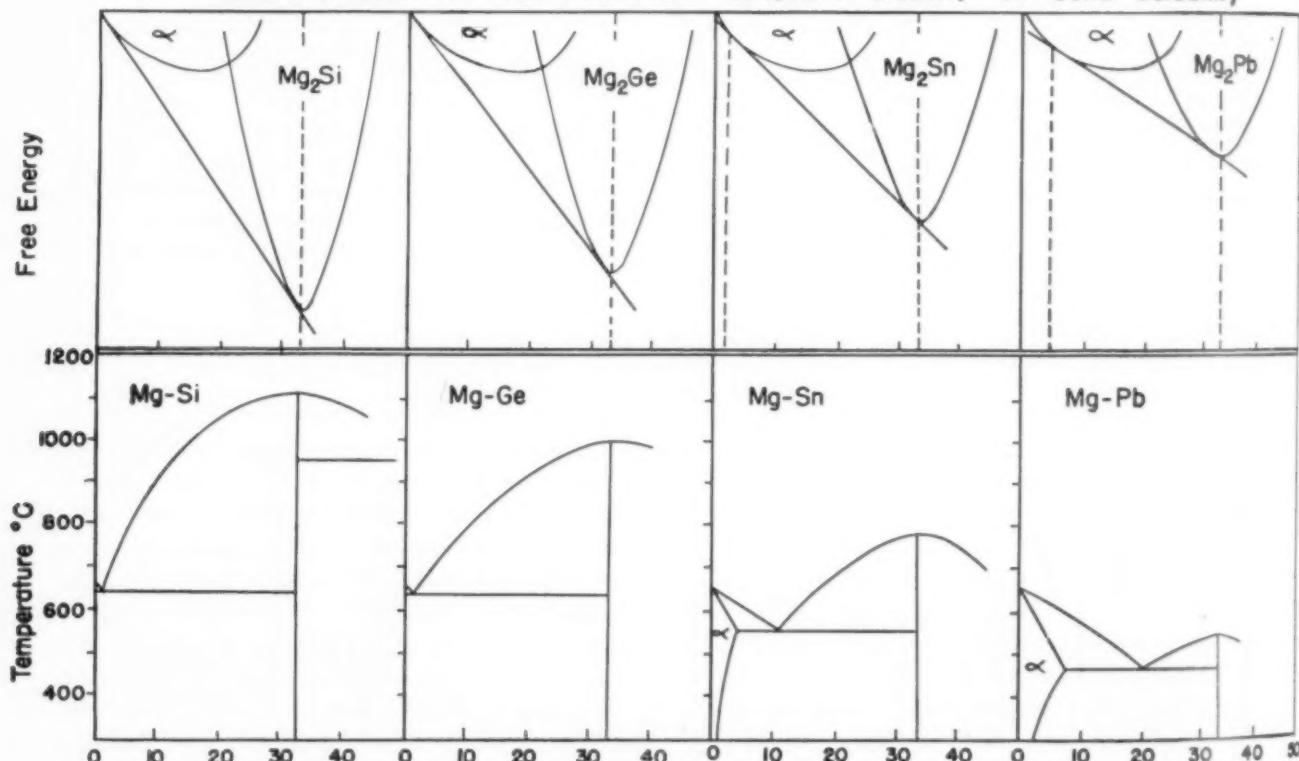
The slope of such a tangent is decreased as the compound stability is lowered, so that the composition at which it touches the free energy curve of the  $\alpha$ -phase is progressively increased. In alloy systems with metals of comparatively high melting point (metals from the transition series, that is to say) where the tendency for compound formation is not very prominent, the alloying components produce the same influence on the  $\alpha$ -phase as a compound with high melting point, because of their relatively high thermody-

namic stabilities. The resulting effect is therefore governed by the free energy curves of the phases under consideration.

**Effect of Compound Stability**—The general influence of compound stability on the maximum solubility of metals in magnesium may be seen in Fig. 6. The solubility becomes rapidly less as the melting point of compounds increases. The reason for the greater solubility of silver in magnesium than gold, even though their atomic sizes and valence factors are identical, is explained in Fig. 6 by the fact that the electrochemical factor of gold is more effective than that of silver.

A considerable influence is also exerted by compound stability on both eutectic composition and eutectic temperature, as illustrated in Fig. 7 and 8, respectively (center of page 303). In general, a rapid decrease in the eutectic composition and a gradual increase in the eutectic temperature are experienced as the melting points of the compounds are increased. There also appears to be a direct relationship between the eutectic composition and the eutectic temperature, as brought out in Fig. 9 (top right). The curve, moreover, extrapolates to 650° C., a temperature

Figure 5. Influence of Size Factor and Compound Stability on Solid Solubility



Supplementary Data:	
1. Crystal Structure	---
2. Size factor	---
3. Alloying Valence	---
4. Group	IV b
5. Period	3

A. Silicon

Dia. Cubic  
-26% (unfavorable)  
4  
IV b  
3

B. Germanium

Dia. Cubic  
-23% (unfavorable)  
4  
IV b  
4

C. Tin

BCT  
-12% (favorable)  
4  
IV b  
5

D. Lead

FCC  
+9% (favorable)  
4  
IV b  
6

exactly equal to the melting point of magnesium. Further attention to binary compounds of magnesium indicates (as presented in Fig. 10 on the next page) that their relative stabilities depend upon the position of the metals in the periodic system. The curves for the four elemental periods are somewhat parabolic in form. With only one exception, that is, period 5, all curves have a minimum for metals in Group III-b. Evidently the compounds from III-b are least stable and have about the same melting point, irrespective of the period to which they belong. Stability of compounds increases rapidly in any one period the further the position

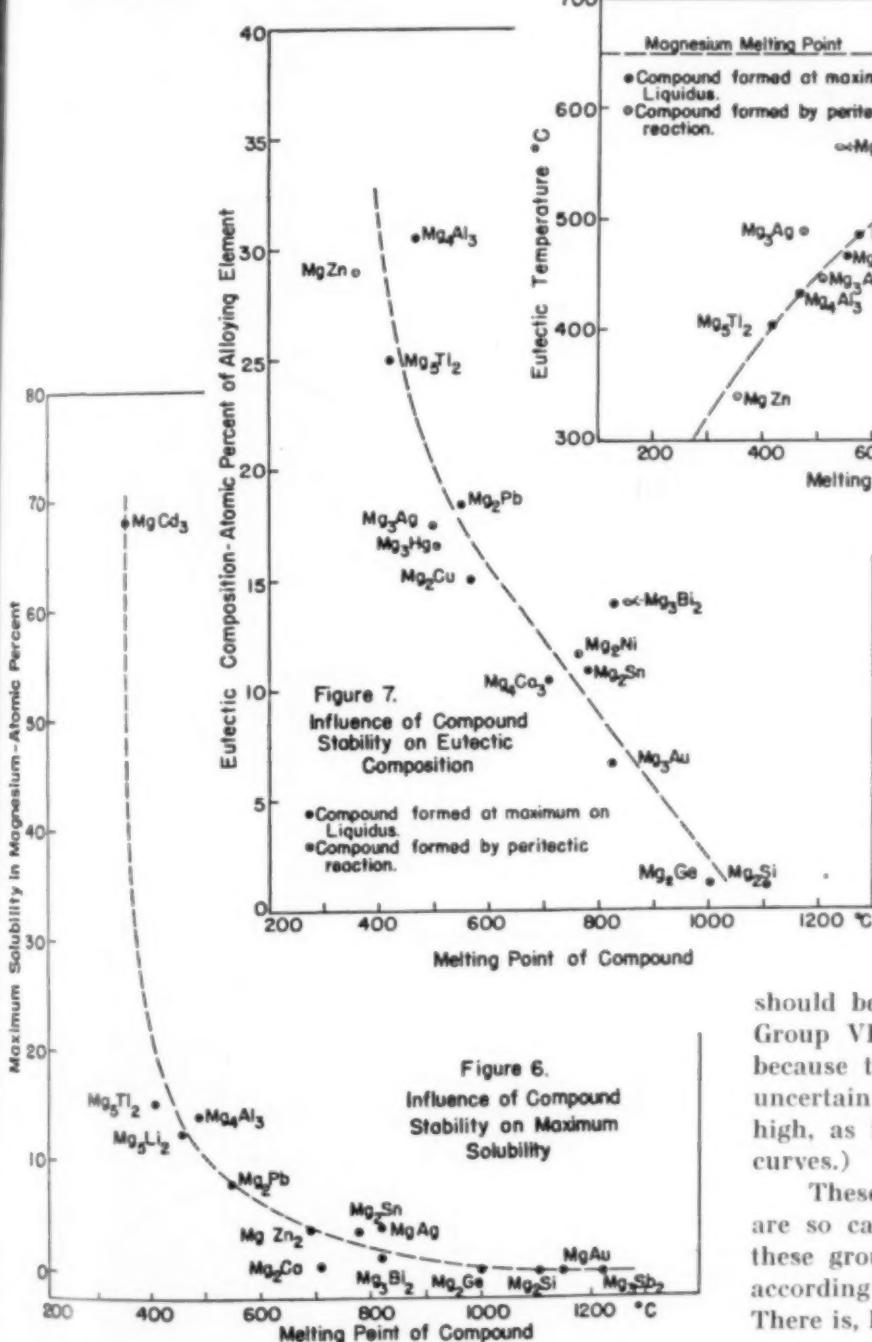


Figure 6.  
Influence of Compound  
Stability on Maximum  
Solubility

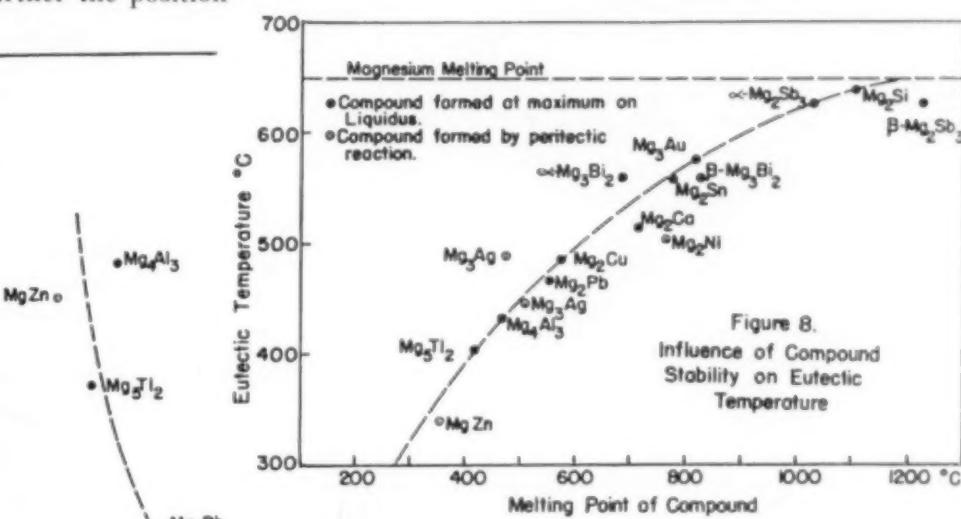


Figure 7.  
Influence of Compound  
Stability on Eutectic  
Composition

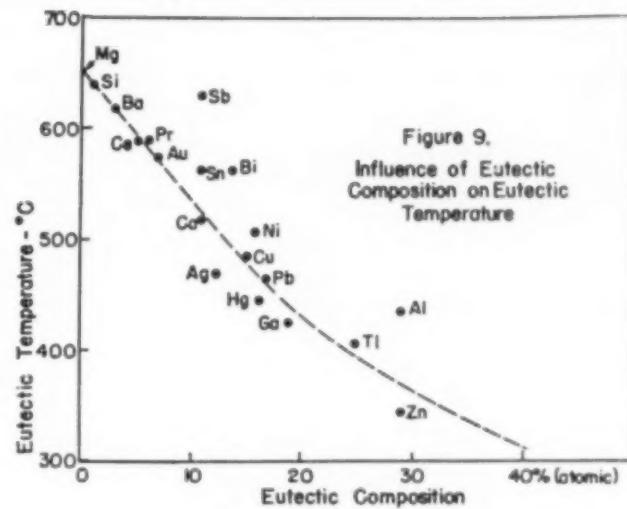


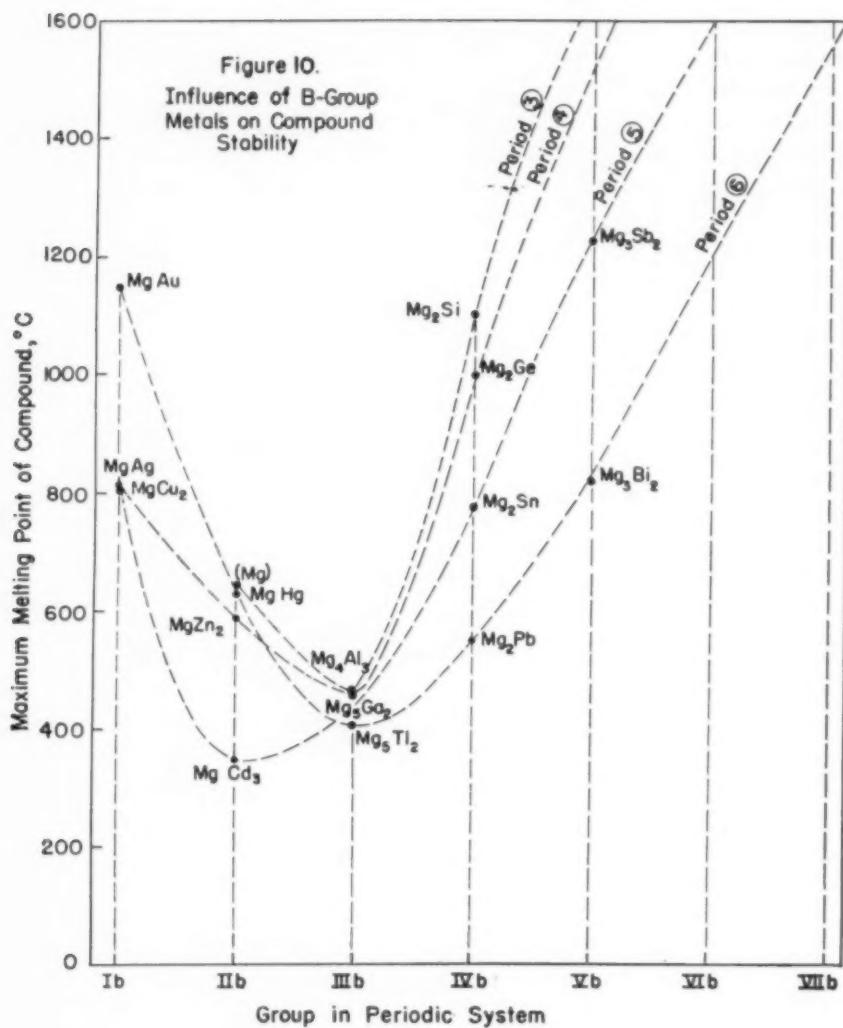
Figure 8.  
Influence of Compound  
Stability on Eutectic  
Temperature

of the compound is from Group III-b.

**Normal Valence Compounds**—Moreover, there is a strong tendency for magnesium to form normal valence compounds with metals from Groups IV-b to VI-b of the following types: (a) Mg<sub>2</sub>(Si, Ge, Sn, or Pb) in Group IV-b; (b) Mg<sub>3</sub>(As, Sb, or Bi)<sub>2</sub> in Group V-b; (c) Mg(S, Se, or Te) in Group VI-b. (It

should be noted that compounds from Group VI-b are not shown in Fig. 10 because their exact melting points are uncertain—although claimed to be very high, as indicated by the trend of the curves.)

These "normal" valence compounds are so called because the metals from these groups combine with magnesium according to their chemical valence. There is, however, no tendency for mag-



nesium to form normal valence compounds with metals in other groups. Instead, typical intermetallic compounds are produced.

Finally, as a matter of precaution, it is necessary to warn that *quite explosive reactions result in the formation of certain compounds of magnesium* such as  $MgSe$  and  $MgTe$ , in which the reactions occur with a violent evolution of heat. The compounds  $Mg_3As_2$ ,  $Mg_3Bi_2$ , and  $Mg_3Sb_2$  also form spontaneously and are accompanied by the release of considerable heat; reactions forming these compounds may also become highly explosive if care is not exercised.

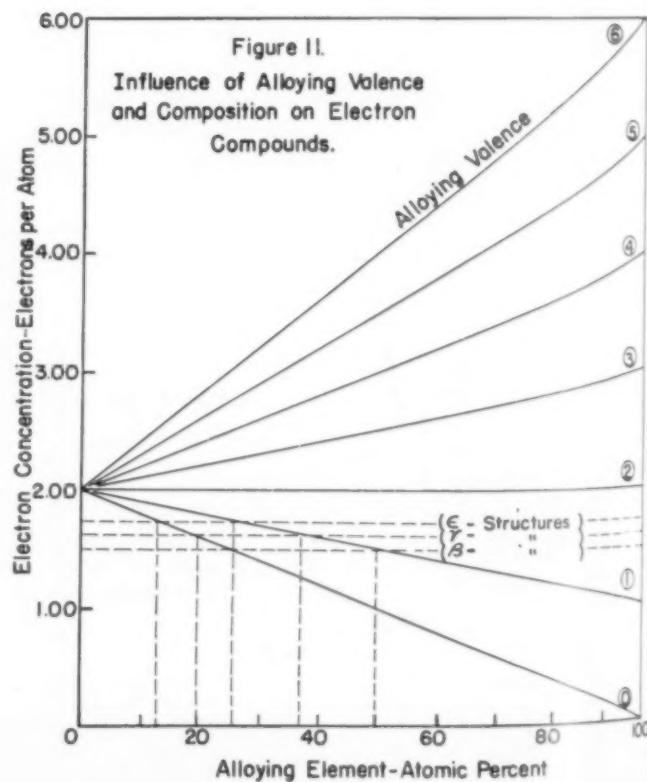
#### Miscellaneous Alloying Behavior

As magnesium under special conditions may exhibit alloying conduct differing from the deportment already discussed, this unusual behavior will be touched upon in order to define fully the alloying nature of the metal magnesium.

**Electron Compounds** — It has been shown

that intermetallic compounds will form wherever the electrochemical factors are prominently active. But other types of alloy compounds may occur even though the electrochemical factors are none too decisive. These products are known as electron compounds because their crystal structures are essentially determined by the number of free or valence electrons contributed by the alloying atoms. Electron compounds are actually phases with wide ranges of homogeneity, in contrast to the restricted ranges exhibited by general intermetallic and normal valence compounds.

That structure and electron concentration are related was first concluded from experimental evidence and later confirmed by theory. It was established that cubic structures of either the  $\beta$ -brass or the  $\beta$ -manganese type occur at electron:atom ratios of 3:2, complex cubic structures of the  $\gamma$ -brass type at 21:13, and the hexagonal close-packed structures of the  $\epsilon$ -brass type at 7:4. Moreover, it seems probable that a favorable atomic size factor is



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essential if the composition of such a compound is to vary appreciably.

In a graphical treatment of the electron concentrations of various systems of elements, it is evident, as submitted in Fig. 11, that electron compounds may form only in alloy systems of magnesium with either transition or univalent metals. Then, too, the compositions necessary to produce the proper electron concentrations for the corresponding structures increase in changing from transition to univalent metals. Because these structures may exhibit composition ranges of varying widths (often 10 or 20 atomic per cent), the ratios 3:2, 21:13, and 7:4 are not always included within their respective ranges as found by experiment.

atoms in a random condition, may become ordered at low temperatures if the lattice distortion is high and there is any appreciable atomic mobility. This atomic interchange reduces lattice distortion. In addition, ordering has a profound influence on the thermal, electrical, magnetic, mechanical and chemical properties, and is responsible for many anomalous behaviors.

### Summary and Deductions

Experimental data obtained from numerous alloy systems of magnesium have been considered in this article from the viewpoint of certain empirical and thermodynamical considerations. Several factors, including crystal structure,

Table II — The Relative Influence of Fundamental Factors on the Properties of Alloys

ALLOYING EFFECTS	CRYSTAL STRUCTURE	ATOMIC SIZE FACTOR*	ALLOYING VALENCE FACTOR*	ELECTRO-CHEMICAL FACTOR*	REMARKS
A. Increase in inherent strength, hardness, and ductility of magnesium by means of wide solid solution without heat treatment	Hexagonal	Favorable	Inactive	Inactive	Certain cubic and tetragonal metals may be also employed, to a limited extent
B. Age hardening characteristics	Any	Reasonably large (see Fig. 1)	Active	Active	—
C. Heat treatable properties	Any	Reasonably large (see Fig. 1)	Active	Active	—
D. Improvement of casting properties by reducing a wide solidification range	Any	Favorable	Inactive	Inactive	—
E. Increase in physical properties by controlling crystal structures of alloys	Any	Favorable	Inactive	Inactive	Monovalent or transition metals of proper electron concentrations (see Fig. 11)
F. Improvement of physical properties by order-disorder reactions	Any	Reasonably large (see Fig. 1)	Inactive	Inactive	Atomic ratios of magnesium to solute metal should be either 3:1, 1:1, or 1:3

\*All factors are expressed in a relative sense.

**Order-Disorder** — In certain alloy systems, changes *within* phases which are generally not detectable by microscopic methods have been observed by other means. This phenomenon is the result of the atoms rearranging themselves from a random spatial distribution to an ordered state on a common space-lattice. Such atomic arrangements are called "superlattices" or "ordered structures". Experimental findings make certain that a tendency for ordering exists in systems where atoms are just within the favorable size factor zone and are sufficient in number to form alloys of types such as  $Mg(C')$ ,  $Mg_3(C')$ , or  $Mg(C')_3$ . (In these formulas C' is the alloying component of any given system.)

Alloy systems that are normally stable, with

atomic size, alloying valence, and electrochemical factors of solute metals are found to govern the alloying nature of magnesium. These factors notably affect the liquidus, solidus, and solid solubility curves of equilibrium diagrams. The solidus and solid solubility curves appear to be more susceptible to the action of such factors than the liquidus curves.

Augmented by its relatively high electrochemical nature, magnesium forms eutectiferous types of alloy systems with other metals with few exceptions. The resulting compounds from these systems may exert considerable influence on both the eutectic temperature and composition, and on the limits of solid solubility, depending on their thermodynamical stabilities.

Table III—Supplementary Information on Binary Equilibrium Systems of Magnesium

ELEMENT	DEN-SITY g./cc.	MELT-ING POINT °C.	SEE TABLE I FOR CODE		COM-POUND	MELT-ING POINT OF COM- POUND °C.	EUTEC-TIC TEMPERA-TURE °C.	EUTECTIC COMPOSITION		MAXIMUM SOLUBILITY		ROOM TEMP. SOLUBILITY	
			STRU-C-TURE	SIZE FAC-TOR				ATOMIC %	WEIGHT %	ATOMIC %	WEIGHT %	ATOMIC %	WEIGHT %
Aluminum	2.70	660	FCC	f	Mg <sub>3</sub> Al <sub>2</sub>	463	437	30.0	32.3	11.6	12.2	2.3	2.5
Antimony	6.62	630	Rhomb	f	Mg <sub>3</sub> Sb <sub>2</sub>	1228	629	11.4	39.0	?	?	?	?
Arsenic	5.73	814	Rhomb	unf	Mg <sub>3</sub> As <sub>2</sub>	800?	?	?	?	?	?	—	—
Barium	3.5	704	BCC	unf	Mg <sub>2</sub> Ba <sub>2</sub> +	855	618	2.8	13.8	—	—	—	—
Beryllium	1.85	1284	HCP	unf	?	?	?	?	?	0.14?	0.05?	?	?
Bismuth	9.83	271	Rhomb	f	Mg <sub>3</sub> Bi <sub>2</sub>	823	553	14.3	59.0	1.2	9.1	0.01	0.10
Boron*	2.3	2300	Hex?	unf									
Cadmium	8.65	321	HCP	f	MgCd <sub>3</sub> +	379	—	—	—	68.6?	91.0?	17.8?	50.0?
Calcium	1.55	851	FCC†	unf	Mg <sub>2</sub> Ca	714	517	10.5	16.2	1.1?	1.8?	0.3?	0.5?
Carbon*	2.22	3700?	D-C	unf									
Cerium	6.9	635	HCP†	f	Mg <sub>2</sub> Ce	780	590	4.4	21.0	0.3	1.6	—	—
Cesium*	1.9	28	BCC	unf									
Chromium*	7.14	1860	BCC†	b-l									
Cobalt	8.9	1490	HCP†	unf	?	?	635	2.1	5.0	?	?	—	—
Columbium*	8.57	1950	BCC	f									
Copper	8.94	1083	FCC	unf	MgCu <sub>2</sub> +	819	485	14.5	30.7	0.01	0.03	0.008	0.02
Gallium	5.91	30	Orth	f	Mg <sub>2</sub> Ga <sub>3</sub> +	456	423	19.1	42.9	3.1	8.4	0.3	0.8
Germanium	5.36	958	D-C	unf	Mg <sub>2</sub> Ge	1000	635	1.2	3.5	0.003	0.009	—	—
Gold	19.3	1063	FCC	f	MgAu <sub>2</sub> +	1150	576	7.5	38.0	0.1	0.9	—	—
Indium	7.31	155	FCT	f	MgIn <sub>2</sub> +	?	484	?	?	19.1	52.7	13.0	41.3
Iridium*	4.93	2454	Orth	b-l									
Iron	7.87	1535	BCC	unf	MgFe?	?	?	?	—	—	—	—	—
Lanthanum	6.15	826	HCP	b-l	Mg <sub>2</sub> La	766	571	3.0	15	?	?	?	?
Lead	11.34	327	FCC	f	Mg <sub>2</sub> Pb	550	465	19.1	66.8	7.8	41.8	0.5	4.1
Lithium	0.53	185	BCC	f	Mg <sub>2</sub> Li <sub>2</sub> ?	592	588	21.6	7.3	17.5	5.7	17.5	5.7
Magnesium	1.74	650	HCP	—	—	—	—	—	—	—	—	—	—
Manganese	7.44	1244	BCC†	b-l	?	?	645	?	?	1.5	3.4	—	—
Mercury	13.55	—38.9	Rhomb	f	MgHg <sub>2</sub> +	630	448	18.1	64.5	?	?	?	?
Molybdenum*	10.24	2620	BCC	b-l									
Nickel	8.91	1452	FCC	unf	MgNi <sub>2</sub> +	1145	508	11.3	23.5	0.04	0.1	—	—
Osmium*	22.48	2700	HCP	unf									
Palladium*	12.0	1554	FCC	f									
Phosphorus	1.82	44	Rhomb	unf	Mg <sub>3</sub> P <sub>2</sub>	?	?	?	?	?	?	—	—
Platinum*	21.45	1774	FCC	f									
Praseodymium	6.63	940	HCP	f	Mg <sub>2</sub> Pr <sub>2</sub> +	789	593	4.9	23	?	?	?	?
Rhenium*	20.0	3000	HCP	f									
Rhodium*	12.5	1966	FCC	unf									
Rubidium*	1.53	39	BCC	unf									
Ruthenium*	12.2	2450	HCP	b-l									
Scandium*	2.5	1200	BCC	unf									
Selenium	4.81	220	Hex	unf	MgSe	?	?	?	?	?	?	—	—
Silicon	2.4	1420	D-C	unf	Mg <sub>2</sub> Si	1102	628	1.2	1.4	0.003	0.004	—	—
Silver	10.5	961	FCC	f	MgAg <sub>2</sub> +	820	471	17.6	48.6	3.9	15.3	0.9	3.9
Strontium	2.6	771	FCC	unf	Mg <sub>2</sub> Sr	606	582	5.9	18.4	—	—	—	—
Sulphur	2.07	113	Orth†	unf	MgS	2000?	?	?	?	?	?	—	—
Tantalum*	16.6	2850	BCC	f									
Tellurium	6.24	452	Hex	f	MgTe	?	?	?	?	?	?	?	?
Thallium	11.85	304	HCP†	f	Mg <sub>2</sub> Tl <sub>2</sub> +	413	405	25.0	73.8	15.4	61.7	7.3	39.9
Thorium*	11.5	1850	FCC	f									
Tin	7.30	232	BCT†	f	Mg <sub>2</sub> Tn <sub>2</sub> +	778	561	10.7	36.9	3.4	14.6	0.2	1.0
Titanium*	4.51	1800	HCP†	f	Mg <sub>2</sub> Ti <sub>2</sub> +	?	?	?	?	?	?	?	?
Tungsten	19.3	3370	BCC	f									
Vanadium*	18.7	1715	BCC	unf									
Zinc	7.14	419	HCP	f	MgZn <sub>2</sub> +	590	341	30.4	54	3.3	8.4	0.6	1.5
Zirconium	6.4	1860	HCP†	f	?	?	?	?	?	0.2	0.8	0.04	0.2

\*Data on alloys unavailable.

†Other structures also exist.

+ Other compounds exist at lower temperatures.

Certain magnesium compounds form with explosive reactions. This violence is typical of compounds produced with the evolution of considerable heat owing to the exceptionally high electrochemical factor or thermodynamical stability. Metals with high melting points behave thermodynamically much like highly stable compounds, even though they may have no tendency to form compounds with magnesium; under this condition, no explosive reactions will occur upon alloying, but limited solid solubility can be expected.

Apart from producing compounds of either the normal valence or the general intermetallic type, magnesium may also form electron compounds. The formation of the latter may result only where magnesium alloys with either monovalent metals or transition metals at the proper electron concentrations. Under certain conditions, magnesium may also produce alloys with ordered structures.

By exploiting judiciously the fundamental factors concerning the alloying nature of magnesium, we are approaching the time when alloys may be made with specified properties. The relative influence of the fundamental factors on cer-

tain properties of alloys is given in Table II. By selecting the proper combination of alloying metals, the conditions for obtaining certain desirable physical properties may be satisfied. Conversely, from the knowledge of both the composition and the relative influence of the fundamental factors, it is possible to predict certain physical properties of a given alloy.

#### *Supplemental Information on Binary Equilibrium Systems of Magnesium*

An up-to-date collection of pertinent data on binary systems of magnesium, that is, the compounds and their melting temperatures, eutectic compositions, eutectic temperatures, and limits of solid solubilities, is contained in Table III. Additional information on pure metals and their respective alloy possibilities with magnesium is also included therein.

The reason for appending this table is two-fold: It serves as a compact summary of essential data on all magnesium-rich binary equilibrium diagrams which have been published so far; and it offers suggestions for selecting new binary systems meriting investigation.

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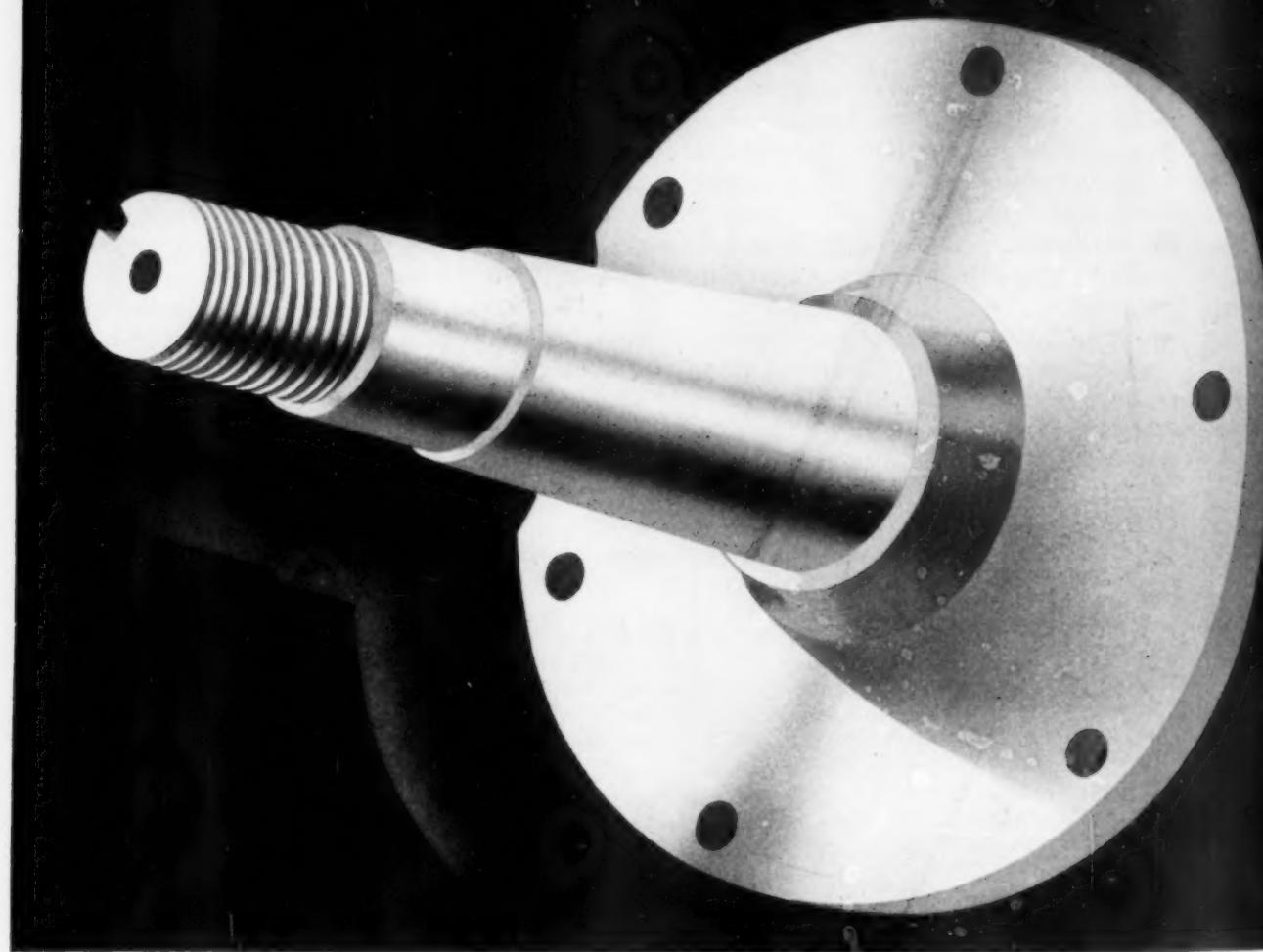
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## Personals

Elected as officers of the Electric Metal Makers Guild, Inc.: J. A. DEBONDY , superintendent of melting, Manitoba Steel Foundries, Ltd., president; J. E. ARTHUR , superintendent of melting, Crucible Steel Co. of America, vice-president; D. L. CLARK , superintendent of melting, Simonds Saw and Steel Co., secretary-treasurer.

Appointments by Vanadium-Alloys Steel Co., Latrobe, Pa.: ROY C. MCKENNA , to president, replacing FLOYD ROSE , resigned; R. BURNS GEORGE , to sales vice-president; L. D. BOWMAN , to production vice-president; F. P. UNDERWOOD to vice-president and secretary; J. P. GILL , vice-president, appointed chairman of a newly formed Executive Committee; J. CLEVELAND MCKENNA , elected a director; LAWRENCE WOOD, appointed advertising manager and

member of the Sales Committee; W. R. MAU , member of the Sales Committee; HENRY HUDSON , Boston manager; RIDGWAY COOK , Springfield, Mass., manager; FRAN MOBERG, works manager of Latrobe plant; HARRY GOOD , works manager of Anchor Drawn Steel Co.; GEORGE A. ROBERTS , chief metallurgist; HARRY G. JOHNSTIN , metallurgical engineer.

DANA W. SMITH , formerly assistant chief of the Metal Working Division, Aluminum Co. of America, New Kensington, Pa., is now principal research metallurgist for the Glenn L. Martin Co., Baltimore, Md.

TRUMAN S. FULLER , formerly assistant engineering head of the General Electric Works Laboratory, Schenectady, has been appointed engineer in charge of the Works Laboratory.

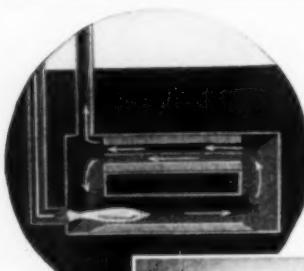
R. J. McCracken , formerly assistant plant manager of the Aviation Division of the Studebaker Corp., Fort Wayne, Ind., has been named general manager of the Durham Mfg. Corp., Fort Wayne Division. JOHN BAILEY , formerly with Peerless Corp. of America in Muncie, has been named foreman of the Durham heat treating department.

ARNOLD P. SEASHOLTZ , has resigned as metallurgical engineer with E. F. Houghton & Co. and is now conducting his own business, Seasholtz Metallurgical Service, in Lancaster, Pa.

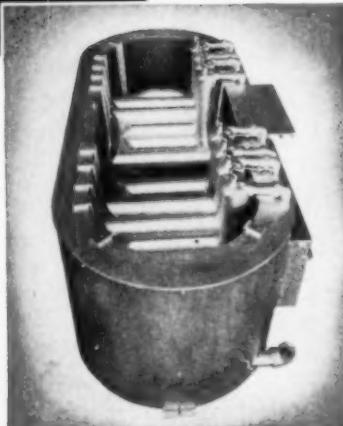
G. L. DENISTON , formerly associated with the Aero Products Division, General Motors Corp., Dayton, Ohio, has been appointed to the staff of Battelle Institute, Columbus, Ohio, as a chemical engineer.

S. L. GERTSMAN , is now associated with the Hull Iron & Steel Foundries Limited, Hull, Canada, as director of process control and research, and is also acting as the liaison officer with the Physical Metallurgy Research Laboratory of the Department of Mines.

S. E. MADDIGAN , formerly with Chase Brass & Copper Co., Waterbury, Conn., is now director, B.C. Industrial & Scientific Research Council, at the University of British Columbia, Vancouver, B.C., Canada.



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# Personals

BOYD E. CASS  $\ominus$ , formerly senior metallurgist at Baldwin Locomotive Works, Eddystone, Pa., is now welding engineer for Foote Mineral Co., Philadelphia.

LILLIAN H. HAMMOND  $\ominus$ , formerly with Hardy Metallurgical Co., has joined the research and development staff at Sherman & Co., New York, as metallurgist.

M. D. JOHNSON  $\ominus$  has resigned as chief inspector at Caterpillar Tractor Co., and has joined the staff of the director of manufacturing of Graham-Paige Motors Corp., Detroit.

LOUIS A. DELAUCHE  $\ominus$  is now with British American Foundry & Machine Limited, St. Thomas, Ont., as assistant manager.

R. A. CLARK  $\ominus$  has severed his connection with the Lakey Foundry & Machine Co., Muskegon, Mich., and is now metallurgist for Electro Metallurgical Co. in Chicago.

EDWARD A. ERICKSON  $\ominus$ , formerly associated with the U. S. Navy Bureau of Ships, Navy Yard, New York, has been named to the staff of Battelle Memorial Institute, Columbus, Ohio.

R. CLARK DAWES  $\ominus$ , formerly assistant professor of chemical engineering and metallurgy at Grove City College, is now research engineer at Battelle Memorial Institute, Columbus, Ohio.

FRANK SPITALE  $\ominus$  has returned to the Beryllium Corp. of Pennsylvania as superintendent of the melting department.

J. F. JOY, formerly director of engineering for the Federal Machine and Welder Co., Warren, Ohio, has been elected vice-president in charge of engineering.

Promoted by Falcon Bronze Co., Youngstown, Ohio: J. C. LOPATTA  $\ominus$  from general foreman to foundry superintendent.

Promoted by American Steel & Wire Co.: WADE B. HOUK  $\ominus$ , from works metallurgist at South Works to assistant superintendent of the Wire Division.

Transferred by Wolverine Tool Co. to Division of Calumet & Hecla Copper Consolidated, Detroit: J. J. LOPRETE  $\ominus$ , from chief chemist in charge of the chemical and physical testing laboratories, to plant metallurgist working on furnace operations and mill problems.

Transferred by Magnaflux Corp. to Magnaflux Corp. from manager of Southeastern Division to manager of West Coast District; Roy C. SCHIEBEL, Jr., from West Coast manager to Chicago headquarters offices; W. D. REID, Jr.  $\ominus$ , from New York to manager of Southeastern Division.

CHARLES S. DUMONT  $\ominus$ , formerly on the superintendent's staff, Electro Metallurgical Co., Niagara Falls, N. Y., is now a research engineer for Battelle Memorial Institute.

RALPH J. EVARTS  $\ominus$  is now a research and development engineer with Santay Corp., Chicago.

J. C. BARRETT  $\ominus$ , formerly metallurgist for the Taylor-Winsfield Corp., is now employed at the Glenn L. Martin Co., Baltimore, as research metallurgist.

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